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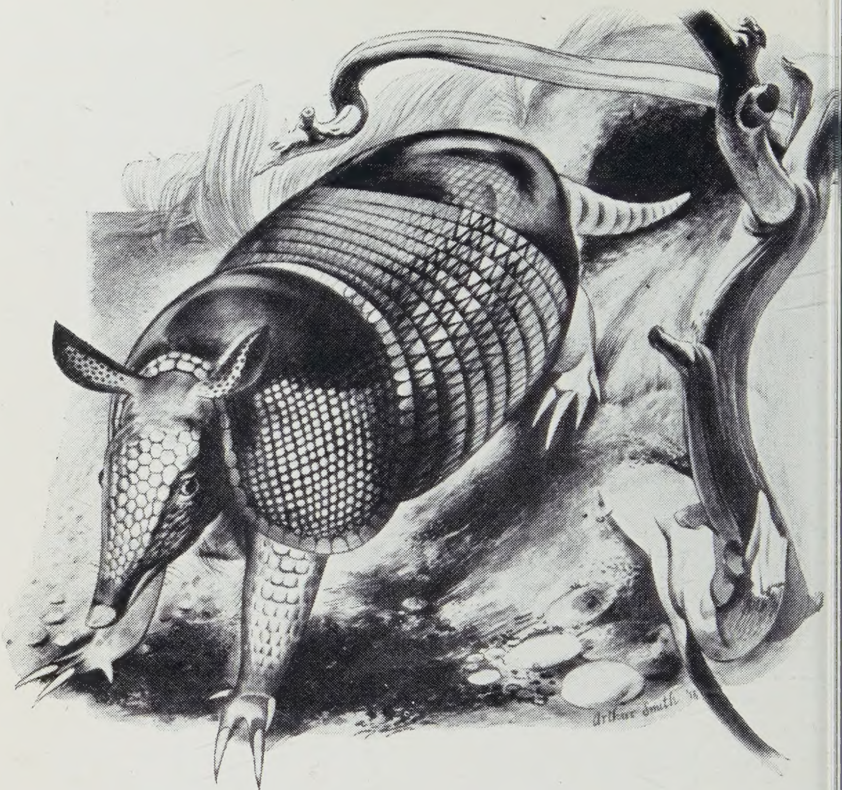
The Journal of the INSTITUTE OF METALS

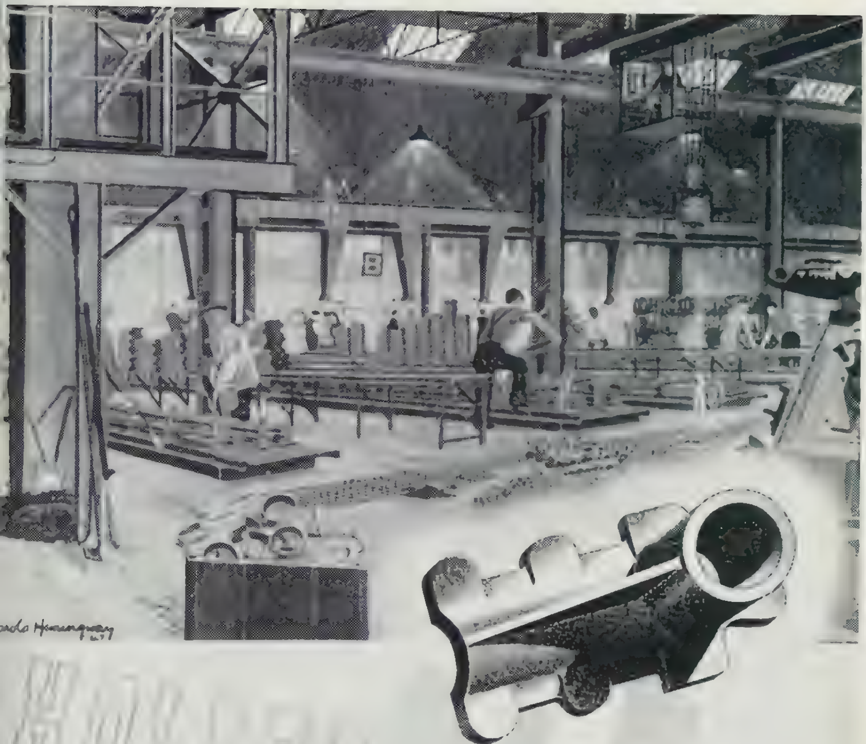
and

METALLURGICAL ABSTRACTS

In this Issue:

JUNE 1948



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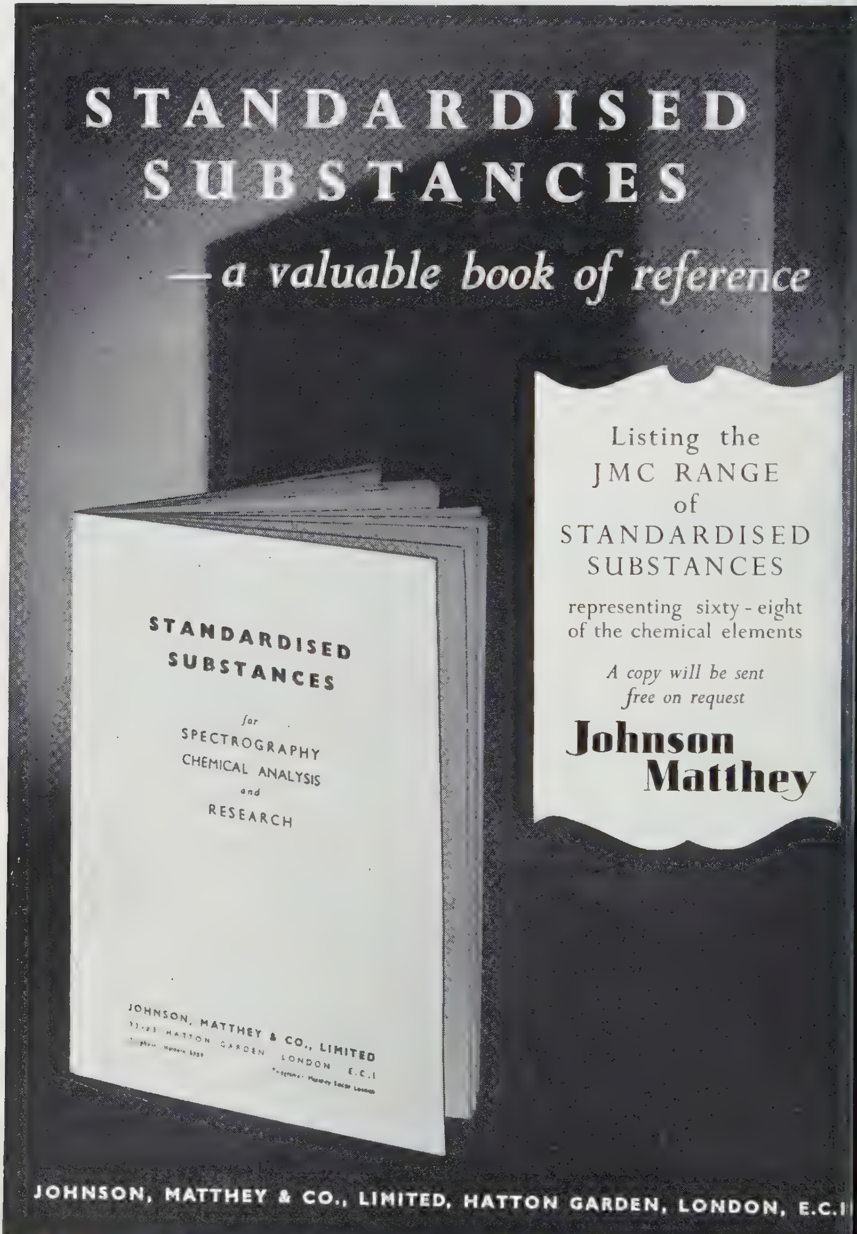
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'HE JOURNAL OF THE INSTITUTE OF METALS

JUNE 1948

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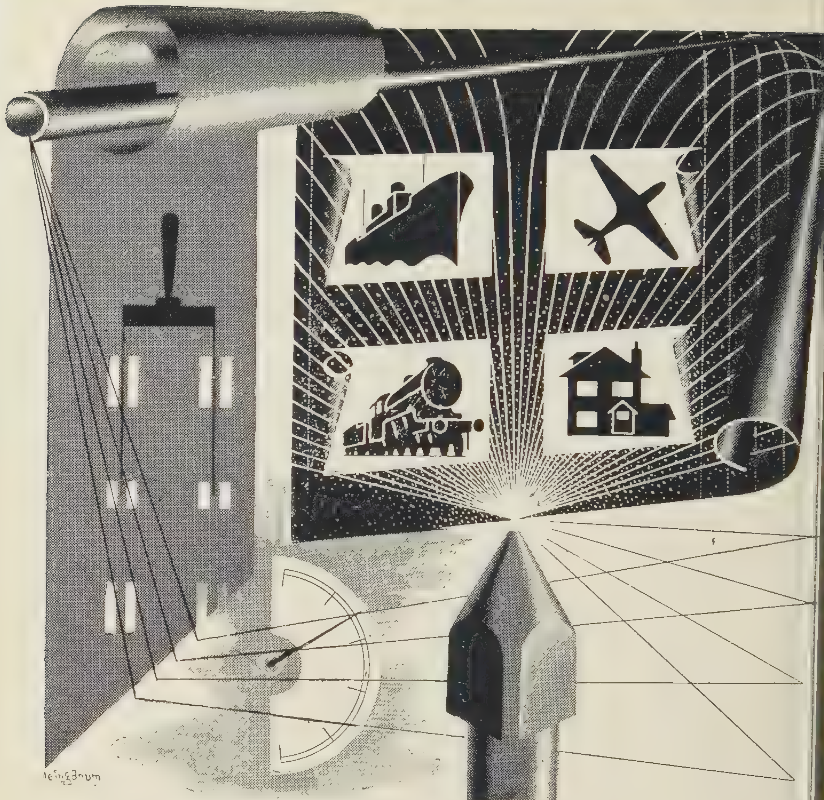

THE INSTITUTE OF METALS

President:
Sir ARTHUR SMOUT, J.P.
Secretary and Editor of Publications:
Lieut.-Colonel S. C. GUILLAN, T.D.

Assistant Editor:
Major W. G. ASKEW, M.C.

Assistant Secretary:
Major R. E. MOORE

Administrative and Editorial Offices:
4 GROSVENOR GARDENS, LONDON, S.W.1

Telephone:
SLOANE 6233

INSTITUTE NEWS AND ANNOUNCEMENTS

AUTUMN MEETING, CAMBRIDGE, 14–17 SEPTEMBER 1948

The Council hopes that there will be a large attendance at the 1948 Autumn Meeting, to be held in Cambridge, for which an attractive programme has been arranged. The programme, and a reply form, will be despatched to all members at an early date.

The Autumn Meetings provide particularly good opportunities for members and their friends to meet socially, and it is expected that this year's meeting at Cambridge will be attended by many Continental and other overseas members.

ANNUAL GENERAL MEETING, 1949

The 1949 Annual General Meeting of the Institute will be held in London on Wednesday, Thursday, and Friday, 30 and 31 March and 1 April 1949.

In connection with this meeting, a Symposium is being arranged on " Metallurgical Aspects of Non-Ferrous Metal Melting and Casting of Ingots for Working ".

A Dinner-Dance is planned to be held on the evening of Thursday, 31 March, at the Savoy Hotel, on the lines of the very successful function held in 1947.

MEMBERSHIP DEVELOPMENT

In their Report for the Year Ended 31 December 1947, published in the February 1948 issue of the *Journal*, the Council stated: " It is becoming generally recognized that the *Journal* and *Metallurgical Abstracts* are essential to every metallurgist engaged in the field of non-ferrous metallurgy. The Council must emphasize, however, that the problem of finance still limits the Institute's services and the development of its publications. While the present membership is not unsatisfactory, it is capable of extensive development in the British Isles and overseas, as it is still by no means fully representative of the field which the Institute serves. It is hoped that members will make considerable efforts during the forthcoming year to increase it. A large increase in membership would enable the Institute to develop its services and, in particular, to improve and extend its publications. A List of Members, which will be issued in the first half of 1948, will be of great assistance in membership development."

As the Institute's financial year commences on 1 July, the present is a particularly suitable time at which to join. Members are, therefore, invited to take energetic steps to increase the Institute's membership among those of their friends who should be interested. Pending the issue of a List of Members (now in the press), the Secretary will be pleased to confirm whether or not persons whom members wish to invite to join the Institute are already members or are about to be elected. He will also be pleased to supply application forms and specimen copies of the monthly *Journal* for membership development purposes.

DINNER TO PROFESSORS A. O. RANKINE AND R. F. MEHL

The President and Council gave a dinner party at the Savoy Hotel, London, on 13 May in honour of Professor A. O. Rankine, O.B.E., D.Sc., F.R.S., 1948 May Lecturer, and Professor R. F. Mehl, Sc.D., Ph.D., Director of the Metals Research Laboratory and Professor of Metallurgy at the Carnegie Institute of Technology, Pittsburgh, Pa., U.S.A.

There were present: Sir Arthur Smout (*President*), in the Chair, Professor W. R. Jones (*President of the Institution of Mining and Metallurgy*), Dr. J. W. Jenkin (*President of the Institution of Metallurgists*), Mr. K. Headlam-Morley (*Secretary, Iron and Steel Institute*), Dr. C. H. Desch, Lieut.-Colonel Sir John Greenly, and Colonel P. G. J. Gueterbock (*Past-Presidents*), Mr. W. F. Brazener, Mr. J. Cartland, Mr. A. J. Murphy, and Mr. H. S. Tasker (*Vice-Presidents*), Mr. W. A. C. Newman (*Honorary Treasurer*), Professor Leslie Aitchison, Major C. J. P. Ball, Mr. D. F. Campbell, Mr. R. Chadwick, Dr. B. Chalmers, Dr. Maurice Cook, Mr. T. M. Herbert, Mr. D. W. Hopkins, Mr. D. P. C. Neave, Professor H. O'Neill, Dr. L. B. Pfeil, Mr. A. R. Powell, and Mr. W. F. Randall (*Members of Council*), Mr. G. L. Bailey (*Honorary Member of Council*), and Lieut.-Colonel S. C. Guillan (*Secretary*).

DINNER TO MR. ROBERT C. STANLEY AND DR. PAUL D. MERICA

The President and Council gave a dinner party at the Savoy Hotel, London, on 26 May, in honour of Mr. Robert C. Stanley, Chairman and President of the International Nickel Company of Canada, Ltd., and Dr. Paul D. Merica, Vice-President of the International Nickel Company.

During the evening, the President presented to Mr. Stanley the Institute of Metals (Platinum) Medal for 1948, awarded to him in recognition of his outstanding services to the non-ferrous metal industries.

Those present were Sir Arthur Smout (*President*), in the Chair, Sir William Griffiths and Dr. Richard Seligman (*Past-Presidents*), Mr. W. F. Brazener, Mr. J. Cartland, and Dr. S. F. Dorey (*Vice-Presidents*), Mr. D. F. Campbell, Dr. Maurice Cook, Mr. D. P. C. Neave, Dr. L. B. Pfeil, Mr. A. R. Powell, Mr. W. F. Randall, and Dr. C. J. Smithells (*Members of Council*), Mr. G. L. Bailey (*Honorary Member of Council*), and Lieut.-Colonel S. C. Guillan (*Secretary*).

Other officers of the Institute were unavoidably prevented from being present, owing to the short notice that could be given.

COMMITTEES FOR 1948–49

The Council has appointed the following committees to serve for the year 1948–49:

FINANCE AND GENERAL PURPOSES COMMITTEE.

Ball, Major C. J. P. (*Chairman*).
Austin, Professor G. Wesley.
Brazener, Mr. W. F.
Greenly, Lieut.-Colonel Sir John.
Griffiths, Sir William.
Gueterbock, Colonel P. G. J.
Jenkin, Dr. J. W.
Murphy, Mr. A. J.
Neave, Mr. D. P. C.
Tasker, Mr. H. S.

Ex-officio :

The President.
Honorary Treasurer.
Chairman, Publication Committee.

LOCAL SECTIONS COMMITTEE.

Brazener, Mr. W. F. (*Chairman*).
Aitchison, Professor Leslie.
Cartland, Mr. J.
O'Neill, Professor H.
Powell, Mr. A. R.
Thompson, Professor F. C.
Chadwick, Mr. R. (*Chairman, Birmingham Local Section*).
Bucknall, Mr. E. H. (*Hon. Secretary, Birmingham Local Section*).
Randall, Mr. W. F. (*Chairman, London Local Section*).
Rhodes, Dr. E. C. (*Hon. Secretary, London Local Section*).
Macdonald, Mr. A. Craig (*Chairman, Scottish Local Section*).
Hay, Mr. Matthew (*Hon. Secretary, Scottish Local Section*).
Orme, Major F. (*Chairman, Sheffield Local Section*).
Maddocks, Dr. W. R. (*Hon. Secretary, Sheffield Local Section*).
Hopkins, Mr. D. W. (*Chairman, South Wales Local Section*).
Spring, Mr. K. M. (*Hon. Secretary, South Wales Local Section*).

Ex-officio :

The President.
Honorary Treasurer.

METAL PHYSICS COMMITTEE.

Allen, Dr. N. P. (*Chairman*).
Bailey, Mr. G. L. (*co-opted*).
Cook, Dr. Maurice.
Hume-Rothery, Dr. W. (*co-opted*).
Murphy, Mr. A. J.
Oliver, Mr. D. A. (*representing Iron and Steel Institute*).
O'Neill, Professor H.
Raynor, Dr. G. V. (*co-opted*).
Smithells, Dr. C. J.

Ex-officio :

The President.
Honorary Treasurer.
Chairman, Publication Committee.

PUBLICATION COMMITTEE.

Pfeil, Dr. L. B. (*Chairman*).
Aitchison, Professor Leslie.
Cook, Dr. Maurice.
Desch, Dr. C. H. (*co-opted*).
Gueterbock, Colonel P. G. J.
Herbert, Mr. T. M.
Macdonald, Mr. A. Craig.
Murphy, Mr. A. J.
O'Neill, Professor H.
Raynor, Dr. G. V. (*co-opted*).
Underwood, Dr. L. R. (*co-opted*).

Ex-officio :

The President.
Honorary Treasurer.
Chairman, Finance and General Purposes Committee.
Chairman, Metal Physics Committee.

BINDING OF THE JOURNAL AND METALLURGICAL ABSTRACTS

Binding cases for the 1946 volumes have been despatched to all members who have requested them. Cases will be sent, free, to all other members who wish to receive them.

Mr. W. A. Newark, 2 Clerkenwell Green, London, E.C.1, will bind the volumes for members at the rates of 4*s*. 9*d*. per volume or 9*s*. 6*d*. per two annual volumes, inclusive of postage to members' addresses, if desired.

Members are reminded that the *papers* contained in the September to December 1946 issues of the *Journal* must be held for binding in the 1947 volume, which closed with the issue for August 1947 and the special Index issue. Binding cases for the 1947 volume are in preparation.

With effect from 1947 the volumes of both the *Journal* and *Metallurgical Abstracts* close with the August issue, with the addition of index numbers.

BACK ISSUES OF METALLURGICAL ABSTRACTS

Those members who do not possess them are advised to purchase the back issues of *Metallurgical Abstracts* (New Series) from 1934, Vol. I, so that they may possess a set of these invaluable reference books to the literature when the next 10-year Index (1934–1943) is issued. Certain of the volumes are now in short supply.

Members may purchase one copy of each bound volume at the special reduced rate of £1 10*s*., post free.

PERSONAL NOTES

Mr. S. L. Archbutt, F.R.I.C., has now retired from the National Physical Laboratory, and has changed his address to Selborne, Chancellors Park, Hassocks, Sussex.

Mr. L. Badone, B.Sc., A.R.S.M., is now employed with Light Alloys, Ltd., of Renfrey, Ont., Canada.

Mr. E. A. Baker has been appointed Managing Director of Haywood Foundries, Ltd., London.

Mr. W. C. Coppins, M.Sc., A.R.I.C., has left Kent Alloys, Ltd., and is now with the Tin Research Institute, Greenford, Middlesex.

Mr. L. H. Davenport has accepted an appointment with the New Zealand Department of Scientific and Industrial Research, Dominion Physical Laboratory, Wellington; he sailed for New Zealand on 21 April.

Mr. John Fry has changed his office address to Prince George's Road, Merton Abbey, London, S.W.19 (Telephone: Mitcham 2041).

Mr. C. E. H. Hallé, A.M.C.T., has been awarded a Nuffield Vacation Scholarship in Extraction Metallurgy, and will be travelling to Canada early in July.

Mr. Max Horton has resigned his Directorship of the Aluminium Bronze Co., Ltd., Walsall, in view of a number of appointments

that he has undertaken, particularly with the Warren Group of Companies, and as Joint Managing Director of Diecastings, Ltd., Highgate Sq., Moseley Rd., Birmingham 12.

PROFESSOR W. R. D. JONES, D.Sc., Professor of Metallurgy at University College, Cardiff, has been appointed to be a University representative on the National Advisory Committee for Industry and Commerce.

MR. IAN MCLEOD has resigned his post with Messrs. Production Tooling Units in order to take up an appointment with Premier Colloid Mills, Ltd., Brettenham House, Strand, London, W.C.2.

MR. NORMAN PARKINSON, M.Sc., A.R.I.C., has resigned his post at the Nelson Research Laboratories of the English Electric Co., Ltd., Stafford, to take up an appointment as Electrometallurgist at the Tin Research Institute, Fraser Rd., Greenford, Middlesex. His new private address is 76 Hayes End Drive, Hayes End, Middlesex.

MR. JOHN RAE, Jr., A.I.M., has recently been appointed a Departmental Director of McKechnie Brothers, Ltd., and is Works' Director of the Company's Birmingham works.

MR. F. G. RATCLIFFE, J.P., Managing Director of J. F. Ratcliffe (Metals), Ltd., was recently appointed to the Board of Broderick Insulated Structures, Ltd., of Woking and London, which is engaged with the fabrication and installation of light-gauge copper roofs.

MR. F. W. ROWE, B.Sc., has been appointed Vice-Chairman of K. and L. Steelfounders, Ltd., Coborn Works, Letchworth, Herts.

DR. GEORGE SACHS, Director, Research Laboratory for Mechanical Metallurgy and Professor of Physical Metallurgy at the Case Institute of Technology, Cleveland, Ohio, U.S.A., has been appointed Director of the National Metallurgical Laboratory, Jamshedpur, India, and will assume his new duties on 1 October 1948.

The National Metallurgical Laboratory, which is one of five new governmental research laboratories recently established by the Indian Council for Research and Development, will cover all aspects of metallurgical research, both fundamental and applied, and will also carry out research work on ores, minerals, and refractories. Close co-operation with the modern research laboratory of the nearby Tata Iron and Steel Company will be established, and special consideration will be given to research on non-ferrous metals, such as copper, aluminium, manganese, zinc, titanium, and beryllium, of which India has abundant ores, and for the production of which large power resources are available. In the initial stages of the work of the laboratory, the special and urgent needs of India will require particular attention to applied, rather than fundamental, research. The N.M.L., therefore, includes a well equipped Technological Section which will permit establishing and operating pilot plants. On long term research of a fundamental nature, the various National Research Laboratories will work in close co-operation. It is planned at present to provide 35,200 sq. ft. of laboratories. Ample space is also available

for future extension. The main building will have a frontage of 480 ft.

Before joining the Case Faculty in 1939, Dr. Sachs was directing research and teaching in Germany, and has held several executive industrial positions in Germany and the United States. His research work in various fields of metallurgy is generally recognized. In pre-Nazi Germany he was a Fellow of the Kaiser-Wilhelm Gesellschaft, and was the 1941 Annual Lecturer of the Institute of Metals Division, American Institute of Mining and Metallurgical Engineers. At the Case Institute of Technology, Dr. Sachs has taught courses and directed graduate thesis work. His main efforts, however, were aimed at the development of fundamental and applied research. He has also acted as consultant to the International Nickel Company, Chase Brass and Copper Company, Babcock & Wilcox Tube Company, Standard Oil Company of Ohio, and other firms.

Dr. I. G. Slater, M.Sc., F.I.M., has returned from the British Admiralty Delegation, Washington, and has been appointed Director of Operational Research at the Admiralty.

Dr. N. Swindells, B.A., has left the Pressed Steel Company, Ltd., Oxford, and joined McKechnie Brothers, Ltd., Rotton Park Street, Birmingham 16, as Chief Metallurgist.

Colonel T. W. Vigers, O.B.E., M.C., T.D., D.L., has left Germany; his address is 11 Glebe Place, Chelsea, London, S.W.3.

Mr. H. G. Warrington, F.I.M., has left Almin, Ltd., and has taken up a position as Metallurgical Engineer with Dominion Magnesium, Ltd., Suite 406, 67 Yonge St., Toronto 1, Canada.

Mr. L. H. Williams, B.Sc., has changed his address to 4 Aberdeen Rd., Bristol 6.

BIRTHS

Askew.—On 21 May 1948, at Guildford, to Audrey, wife of Major W. G. Askew, M.C., a son.

Moore.—On 25 March 1948, at King's College Hospital, to Carmel, wife of Major R. E. Moore, a daughter (Susan Margaret).

DEATH

The Editor deeply regrets to announce the death in London on 21 May 1948 of Sir William Murray Morrison, Deputy Chairman of The British Aluminium Company, Ltd. Sir William, who was elected a Fellow of the Institute in 1937, served on the Council as a Member of Council from 1912 to 1923, and as a Vice-President from 1923 to 1928. He was an Original Member, and was awarded the 1942 Institute of Metals (Platinum) Medal, in recognition of his outstanding services to the non-ferrous metal industries.

An obituary notice will be published in the *Journal* in due course.

LOCAL SECTIONS NEWS

LOCAL SECTIONS COMMITTEE

A meeting of the Local Sections Committee was held at the offices of the Institute on 12 May 1948, when reports were received regarding the past session and discussions took place on the programmes for the 1948–49 Session. The President of the Leeds Metallurgical Society and the Honorary Secretary of the Manchester Metallurgical Society attended, by invitation.

BIRMINGHAM LOCAL SECTION

The following officers have been elected for the session 1948–49:

Chairman:
R. Chadwick, M.A., F.R.I.C., F.I.M.

Vice-Chairman: B. Thomas, F.Inst.P.

Honorary Secretary: E. H. Bucknall, M.Sc., F.I.M.

Honorary Treasurer: H. H. Symonds, A.I.M.

Representative of the Co-ordinating Committee of the Midland Metallurgical Societies: Maurice Cook, D.Sc., Ph.D., F.I.M.

Past-Chairmen:
N. P. Allen, D.Sc., M.Met., F.I.M.
E. A. Bolton, M.Sc., F.I.M. W. L. Govier, F.I.M.

Ordinary Members:
H. W. G. Hignett, B.Sc., A.R.I.C., F.I.M.
J. O. Hitchcock, B.Sc., F.I.M.
J. W. Jenkin, Ph.D., B.Sc., A.R.I.C., F.I.M.
A. W. Matthews, L.I.M. (representing Student Members).
A. J. Rickard, B.Sc. (representing Student Members).
E. A. Smith (representing Associates).
F. E. Stokeld, F.I.M.

SHEFFIELD LOCAL SECTION

The following have been elected officers for the Session 1948–49:

Chairman:
Major F. Orme, T.D., M.Met., A.R.I.C., F.I.M.

Vice-Chairman: H. G. Dale, F.R.I.C.

Hon. Secretary and Treasurer: W. R. Maddocks, Ph.D., B.Sc.

Past-Chairmen:
Professor J. H. Andrew, D.Sc., F.I.M.
Frank Mason, M.I.E.E.

Ordinary Members:
P. R. Beeley, B.Met., L.I.M. (representing Student Members).
T. B. Bowker.
H. P. Gadsby, Assoc.Met. (representing Associates).
M. M. Hallett, M.Sc., F.I.M.
J. F. B. Jackson, M.Sc., A.R.I.C.
A. J. MacDougall, M.Met.
C. Sykes, D.Sc., Ph.D., F.Inst.P., F.R.S.

OTHER NEWS

MOND NICKEL FELLOWSHIPS

The Mond Nickel Fellowships Committee will shortly be meeting to consider applications for awards of Mond Nickel Fellowships for 1948. Persons wishing to apply for an award who have not sent in completed forms of application should forward them to the Secretary, Mond Nickel Fellowships Committee, 4 Grosvenor Gardens, London, S.W.1, without delay.

DEVELOPMENT COUNCIL FOR THE JEWELLERY AND SILVERWARE INDUSTRY

The Board of Trade have sent to the trade associations and unions in the Jewellery and Silverware Industry their proposals for the establishment of a Development Council under the Industrial Organisation and Development Act, 1947. The organization, which will be called the Jewellery and Silverware Development Council, will be a tri-partite body of employers, workers, and independent members.

The proposals provide for registration of manufacturers, the collection of information and a compulsory levy which, it is suggested, might provide an annual income of about £50,000.

These proposals are tentative, and will be subject to discussion with both sides of the industry. The establishment of development councils is also subject to Parliamentary approval.

BRITISH WELDING RESEARCH ASSOCIATION

The British Welding Research Association in 1945 purchased Abington Hall, an estate near Cambridge, in order that it might be developed as a research station for the engineering work of the Association, and to be complementary to the metallurgical laboratories at the Headquarters in London. On the afternoon of 11 May 1948 the first meeting at Abington Hall of members of firms and organizations of the B.W.R.A. was held. The result was a remarkable success; the sound progress made in the development of the research station was apparent to all, and the enjoyment of the visit was enhanced by the pleasant surroundings seen in perfect weather. Many of the original members with knowledge of the difficulties that were to be overcome, appreciated the transformation even more.

Sir William J. Larke (President), Sir Stanley V. Goodall (Chairman), and Dr. H. G. Taylor (Director of Research) received the visitors, and the attendance was almost to the limit of the accommodation.

ASSOCIATION TECHNIQUE DE FONDERIE

The Twenty-Second Annual Congress of the Association Technique de Fonderie will be held on 8 and 9 October 1948. The technical sessions will be held at the Société des Anciens Elèves des Ecoles Nationales d'Arts et Métiers, 9 bis, Avenue d'Iéna, Paris (16e).

Dr. Marcel Ballay, President of the Association, has issued a cordial invitation to members of the Institute of Metals to take part in the Congress, full particulars of which will be available at a later date.

FIRST NATIONAL MECHANICAL HANDLING EXHIBITION AND CONVENTION

The First National Mechanical Handling Exhibition and Convention will be held in the National Hall, Olympia, London, W.14, from 12 to 21 July, 1948.

Members are invited to attend the exhibition and conference on Friday, 16 July, and to be present at the reading of papers on "Mechanical Handling Methods in the Steel Industry", by Mr. G. W. Grossmith, M.I.Mech.E., at 11.30 a.m., and on "Mechanization in Foundries", by Mr. J. Bain, A.M.I.Mech.E., at 3.30 p.m. Tickets of admission for this date may be obtained on application to the Editor, *Mechanical Handling*, Dorset House, Stamford St., London, S.E.1. The seating accommodation for the convention is limited.

BRITISH ASSOCIATION

The British Association will meet in Brighton from 8 to 15 September 1948. Full particulars may be obtained from the Secretary, British Association, Burlington House, London, W.1.

ROYAL PHOTOGRAPHIC SOCIETY'S 93rd ANNUAL EXHIBITION

The Royal Photographic Society's 93rd Annual Exhibition is to be held at the Society's House, 16 Princes Gate, London, S.W.7, during September and October this year. It will be open in two parts, the first devoted to Pictorial and Colour Photography and the second to Scientific, Technical, Nature, and Record Photography. Lantern slides, colour transparencies and stereoscopic exhibits will be on view throughout the whole Exhibition.

The Scientific and Technical Group of the Society is very anxious to make the second part of the Exhibition fully representative of the best work in all branches of modern applied photography. It feels that it may not be sufficiently widely known that the Exhibition is open to all users of photography and can assist them by giving them a standard against which to judge their own work.

The Scientific Section of the Exhibition is divided, for convenience, into seven classes: Photomicrographs; Radiographs (Medical and Industrial being judged separately); Astronomical and Meteorological; Geological and Metallurgical; Survey (Aerial and Terrestrial); Medical and Surgical; other Scientific Photographs. Nature Photography, Record and Technical Photography (Architectural, Press, Commercial, and Advertising) and Kinematography are in separate Sections. Scientific and technical films should be entered in the last section.

Photographs produced in the course of scientific research are particularly welcomed, especially if they illustrate the use of photography for obtaining information which could not be secured in any other way. The entries must be mounted, and it is urged that a full description be given on the mount, so that the purpose of the work and the value of the results will be clear to all; a series of pictures can be mounted on one card if desired.

Two awards are made annually by the Society in the Scientific Section, namely, the *Hood Medal* for meritorious performance in any branch of practical photography, and the *Rodman Medal* for

outstanding work, preferably in photomicrography or radiography.

Further details will be sent, with the appropriate Entry Form, if application is made to the Secretary, The Royal Photographic Society, 16 Princes Gate, London, S.W.7. There is no entry fee. The closing date for receiving entries is *24 July 1948*.

UNITED NATIONS SCIENTIFIC CONFERENCE ON THE CONSERVATION AND UTILIZATION OF RESOURCES (U.N.S.C.C.U.R.)

The United Nations Scientific Conference on the Conservation and Utilization of Resources, the Provisional Programme of which has been issued, will be held in the United States in May–June 1949, under the authority of the Economic and Social Council of the United Nations.

This Conference represents a new approach to one of the objectives of the United Nations, the promotion of economic development and higher standards of living throughout the world, by a free exchange of ideas and experience on the conservation and utilization of resources. In its plenary meetings, comprising approximately half of the programme, the Conference will examine the contribution of improved techniques to the alleviation of shortages of mineral resources, fuel and energy, and food and forest products. A series of plenary meetings will be devoted to questions of particular importance to under-developed countries.

Specific problems and techniques will be placed on the agenda of specialized section meetings which form the other half of the programme. In these meetings, specialists will discuss Conference Papers on subjects which include fuel and energy, power, and mineral resources.

The Conference will be scientific rather than policy-making. Technical developments will be selected for discussion according to the importance of their practical contribution, actual and potential, to economic development and will be treated so as to interest the economist as well as the scientist, administrator, and engineer.

Conference Papers submitted by experts, as well as summary records of discussions, will be published after the Conference, and it is expected that they will be a continuing and important contribution of the United Nations to better conservation and utilization of resources.

Copies of periodicals containing comments on the provisional programme of the Conference would be appreciated by the Secretariat of U.N.S.C.C.U.R., Department of Economic Affairs, United Nations Secretariat, Lake Success, New York, U.S.A.

ZINC DEVELOPMENT ASSOCIATION

The Association has published its Annual Report for 1947, copies of which may be obtained, on request, from the Secretary, Zinc Development Association, Lincoln House, Turl Street, Oxford.

The Z.D.A. is now entering the tenth year of its existence and the information services which it and its affiliated associations provide to all zinc users are now too well known to call for any particular account of them. This year, therefore, the aim has

been rather to tell the story of zinc in 1947. With a consumption figure of 220,000 tons of virgin zinc a year in the U.K.—the highest level ever recorded in peace-time—zinc is playing a very important part in the work of national recovery; and nowhere is this more evident than in the zinc alloy die-casting industry. While other and bigger uses of the metal, such as for galvanizing and brass manufacture, have continued to increase, a production of 20 million die-castings every week makes a powerful contribution to the export drive in the very wide range of modern goods which contain die-cast components. Recent trends and developments in these and other uses of zinc are all described in the earlier pages of the Report, which then gives an account of the Association's latest publications and various activities in many fields.

The statistical tables on the production of zinc within the Empire, shipments to the U.K., and home consumption are a new feature which show the national significance of the zinc industry.

SCIENCE MUSEUM LIBRARY PHOTOSTAT SERVICE

The Science Museum Library photostat service has now started. Particulars may be obtained from the Science Museum, South Kensington, London, S.W.7.

APPOINTMENTS VACANT

THE CIVIL SERVICE COMMISSIONERS invite applications for the post of HEAD OF THE METALLURGY DIVISION AT THE ATOMIC ENERGY RESEARCH ESTABLISHMENT.

Candidates must be British subjects born on or before 1 August 1917, and should have had extensive post-graduate experience in metallurgical research. Ability to lead a research team is necessary.

The work of the Division includes research on the metallic materials and ceramics of interest in the production and application of atomic energy, and its scope is such as to include fundamental research and the development of technical processes.

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APPOINTMENTS REQUIRED

GRADUATE METALLURGIST (Member of Institute of Metals), Polish nationality, now in this country after war service, with good experience in metal testing and heat-treatment, and as a University lecturer, seeks appointment research work or industrial testing of non-ferrous metals or steel. Please reply Box 215, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

METALLURGIST (age 40), A.I.M. Technical adviser and investigator. Research and development executive. Specialist in powder metallurgy and light alloy application. Box No. 235, Institute of Metals, 4 Grosvenor Gardens, London, S.W.1.

MICRO-HARDNESS TESTING OF METALS.* 1131

By E. WILFRED TAYLOR,† C.B.E., F.Inst.P.

SYNOPSIS.

Reference is made to the methods normally employed for testing the hardness of metals and alloys, and attention is directed to the fact that when an indenter is used the impressions made are usually large in relation to the microstructure of the specimen. The desirability is stressed of a form of hardness test which can be applied to a particular crystal or to a small selected area, and the factors controlling the application of very light loads to a diamond indenter are examined.

A new micro-hardness tester is described, the design of which closely follows the principles discussed earlier. A number of examples are given of micro-hardness tests variously applied to small selected areas of particular specimens.

I.—INTRODUCTION.

MANY well known methods are employed for the routine measurement of the hardness of metals used in the workshop, and all depend on small impressions formed in the surface of the material by an indenter subjected to a known load. Of these methods, the best known are the Brinell, Vickers, and Rockwell, which make use of a steel ball or a diamond pyramid as indenter and apply a known load of 1–100 kg. The results satisfy all ordinary industrial requirements, but the hardness figures so obtained usually represent an average, since nearly all the metallic alloys in common use are far from homogeneous and consist of a mosaic of minute crystals usually much smaller than the indentations made under the loads normally applied.

To those engaged in a detailed study of alloys, some knowledge of the physical properties of the individual crystals and aggregates is essential, and this implies the ability to reveal the boundaries by a suitable etching technique and to apply a hardness test to minute areas present in the specimen. The mosaic formed by a number of components with differing physical properties may be on a very small scale, as, for instance, in hardened beryllium bronze and some steels, so that the etched surface must often be examined under considerable magnification in order to recognize and select appropriate areas for test. Having selected the area, the indenter must be positioned with sufficient accuracy to make the impression precisely at the pre determined point under a light but accurately controlled load. The next

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† Joint Managing Director, Cooke, Troughton, and Simms, Ltd., York.

step necessitates the evaluation of the dimensions of the impression and the assignment of a hardness figure.

Once the means for determining the hardness figures of the components of a microstructure are available, many new applications will be found, as, for instance, the change in hardness as a brazed, soldered, or welded joint is traversed or the depth of the "case" of a case-hardened steel. Similarly, the hardness figures of foils, lamp filaments, or protective plating may be determined without difficulty and the limits of diffusion of one metal into another investigated.

The test may be qualitative and take the form of a scratch made across a prepared surface under carefully controlled conditions, or quantitative with a static indenter acting normally to the surface of the specimen; in the latter case a hardness number may be deduced with considerable accuracy.

II.—General Principles.

Having briefly stated the problem, we may now consider some of the factors involved. In the first place the specimen must be carefully selected, and a face must be ground and polished until it presents a plane surface. This should be accomplished without the generation of sufficient heat to affect the properties of the superficial layer. The specimen must then be treated with a suitable etching solution, the purpose of which is to expose the boundaries between the several components until the latter may be visually recognized. In some cases an electrolytic etching process is preferred, as the surface layer of the specimen, which may be superficially work-hardened by the grinding and polishing technique, is removed in the process. For micro-hardness testing it will be most convenient if the specimen is embedded in a synthetic-resin or plaster mould designed to fit the particular specimen holder, in which case all subsequent mounting difficulties will disappear.

The microscope should be of robust construction and preferably of the inverted or Le Chatelier type, fitted with a mechanical stage to which the micro-hardness testing mechanism can be attached. The eyepiece must carry adjustable cross-wires, and measuring scales can also be incorporated. The indenter may be mounted in the front lens of the objective, or it may form an interchangeable unit which registers exactly with the latter, since it is essential that the impression should be formed precisely in that part of the specimen which is imaged on the cross-wires of the eyepiece. There is much to be said for the former arrangement, since, although the central image-forming rays are occulted, the optical performance of the objective is still sufficiently good to

enable the grain boundaries to be recognized, and the interchange of parts which would have to register with the utmost precision is avoided.

The indenter itself might consist of a minute ball of hardened steel, loaded, as is usual, until a depression is made corresponding to about three-eighths of the diameter of the sphere, but there is much to be said for the selection of the Vickers Diamond Pyramid, which has the faces worked to an angle of 136°. By this choice, elastic deformation of the indenter is avoided, and the hardness figures obtained correspond very closely indeed to those obtained by the Vickers or Brinell methods, without the necessity of choosing the load to suit the ball diameter, or *vice versa*. In short, the same diamond can be used under a load applicable to the particular specimen, and this is a most important consideration.

The load must be accurately applied and may vary from 1·0 to 500 g. A set of weights, such as is usually supplied with a chemical balance, is most convenient and much more reliable in use than any spring loading device. An almost ideal arrangement is one in which the specimen is secured to one end of a carefully balanced lever rocking about almost frictionless bearings. The selected weight is then applied to a platform immediately above the specimen, and the combined objective and diamond indenter are slowly advanced towards the specimen by means of the normal slow-motion mechanism until contact is indicated by some visual or aural device. When the loads are very light and the indentation very minute, it is inadvisable to reverse the fine motion while the diamond is in contact with the specimen, as all mechanisms are subject to some degree of hysteresis, and it is most important that no further movement of the indenter should take place. With light loads of 20 g. or under, a safer alternative is to raise the specimen gently out of contact with the diamond. Having examined some of the problems associated with the design of a micro-hardness tester, an instrument will be described in which an attempt has been made to secure the most favourable working conditions.

III.—Apparatus.

The following is a description of the apparatus as finally developed.

Referring to Fig. 1 (Plate XXXII), a base casting *A* is secured to the stage of the microscope by means of two knurled heads *B*. To this base a lever *C* is pivoted in ball races at *D*. At one end of the lever *C* are removable weights *E* and an adjustable weight *F*, and at the other end are a circular platform *G* (on which weights may be placed) and a removable specimen holder *H* which registers in a dove-tail slide and

is clamped by the finger screw J. The lever C may be secured to the base-plate A by means of the finger screw K, but on the release of this the conductor L will, on depression of this end of the lever, make contact with a second insulated conductor M and permit current to flow through a lamp housed in N, with the consequent illumination of a red glass window O.

The supply current at 4 V. is brought to terminals P and Q (not shown) on the base-plate A, and the objective in which the diamond indenter is mounted is at R.

The micro-hardness tester is shown in Fig. 1 mounted on the stage of a standard Vickers Projection Microscope on which the traversing screws are at U.

The sequence of operations is as follows:

(1) The specimen holder H is released by means of the finger screw J and removed for the insertion of the specimen. It is then replaced and the finger screw J tightened.

(2) The locking screw K is released and the lever C is balanced by the removal or addition of weights at E and the adjustment of the counterpoise F. The flickering of the red light at O is of considerable assistance while balancing.

(3) The selected load is placed centrally on the circular platform G.

(4) The fine-motion head V is used to raise the objective R until the diamond makes contact with the specimen and lifts it sufficiently to break contact between the conductors L and M, as revealed by the failing of the light at O. (The speed of impact of the diamond should be approximately 10 μ/sec., and contact should be maintained for about 15 sec.)

(5) When the load is light, 20 g. or less, it is advisable to lift the specimen clear of the diamond by means of a gentle downward pull on the handle S.

(6) While the specimen is disengaged from the diamond, the slow motion should be reversed before releasing the handle S and refocusing the specimen.

(7) When the specimen is focused, the impression is seen in the field of view and the cross-wires in the eyepiece are brought into coincidence with the impression on the specimen by means of the finger screws T. Thereafter, on bringing any selected point of the image on to the cross-wires, that point on the specimen will receive the impression when the diamond is brought into contact with the specimen.

(8) If it is desired to make a scratch test, the same procedure is followed until the diamond has made contact with the specimen, as indicated by the failure of the red light. One of the stage traverse

screws U is then utilized to move the specimen relatively to the fixed indenter.

Fig. 2 (Plate XXXII) shows the micro-hardness tester unit in its stowage box.

IV.—Examples.

Having now described the means whereby a minute diamond impression under a known pre-determined load can be placed over a selected point on the specimen, it will be of interest to demonstrate the accuracy with which this can be done and to show that with a homogeneous specimen the hardness figure obtained is independent of the load.

The specimen used for Fig. 3 (Plate XXXIII) was a sample of austenitic stainless steel with one surface polished to a mirror finish. On the surface, lines were ruled with a spacing of 0·02 mm. or 20 μ. The loads applied to the diamond were 50, 40, 20, 10, and 5 g., and in each case an attempt was made to place the impression in the centre of one of the ruled squares. It is easily possible to do this with an error of less than 5 μ.

Stainless steel is not an ideal material for testing the effect of a varying load on the hardness figure derived from a measurement of the length of the diagonal of the impression, but Fig. 4 (Plate XXXIII) shows the results obtained with the same specimen. In this case, the loads were 200, 100, 50, 20, 10, 5, 2, and 1 g. The Vickers hardness number V may be derived from the formula:

$$V = \frac{1854 \times L}{d^2}$$

in which L is the load in grammes and d is the average length of the two diagonals of the indentation measured in μ. The hardness figures so obtained are plotted in Fig. 5 A, and those derived from a specimen of phosphor bronze are given in Fig. 5 B.

Grey cast iron contains graphite inclusions, and may consist also of austenite, cementite, and ferrite separately or combined; varying degrees of hardness characterize these different constituents. The conglomerate of ferrite and cementite which constitutes the pearlite eutectoid is easily recognized by its laminated appearance. Fig. 6 (Plate XXXIII) is an example of grey cast iron, showing graphite inclusions, dark areas of pearlite, and light areas of ferrite. The hardness figures are respectively 11, 212, and 266. The load was 20 g. Fig. 7 (Plate XXXIV) is an example of a chilled cast iron, showing dark areas of pearlite and light areas of cementite. In this case the hardness figures

are 245 and 612; the load was 50 g. Fig. 9 (Plate XXXIV) shows a scratch test on the same specimen. Under a load of 0·5 g. the diamond appears to skate over the cementite but to bite into the pearlitic areas. As the load is increased to 1 g. and later to 2 g., the scratch becomes continuous but still reveals the difference in hardness between the light and dark areas.

Fig. 8 (Plate XXXIV) is of Vickers "A" steel (composition: carbon 0·45–0·55, silicon 0·15–0·30, manganese 0·30–0·80, nickel 0·50% maximum) in the unhardened condition, and the impressions were made under a load of 10 g., giving a fairly uniform hardness figure of 200. It is possible that the yielding of ferrite underneath pearlite has

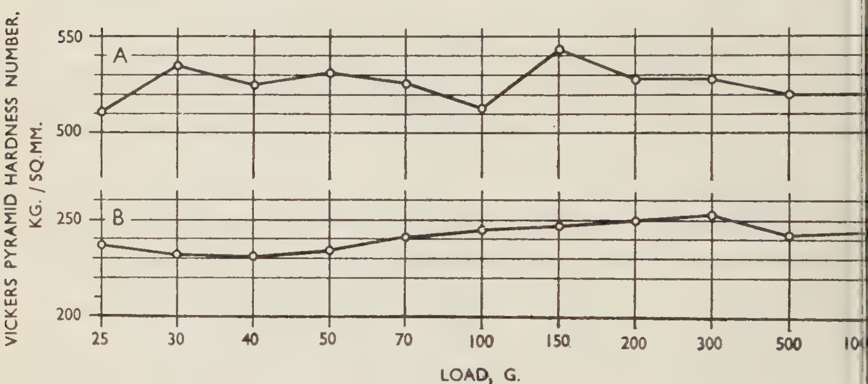

FIG. 5.—A. Ball-Bearing-Quality Stainless Steel (B.S. En. 56).
B. Phosphor Bronze.

given the effect of a spurious softness. Fig. 10 (Plate XXXIV) is of greater interest as showing a section through a piece of Nitralloy steel, chemically hardened from the edge. A series of impressions were made under a load of 50 g., and the hardness figures increase successively from 275 to 930. Fig. 11 shows the change in the hardness number plotted against the distance from the edge measured in microns, and it will be seen that in this instance the hardness figure drops below 900 at about 0·04 mm. from the edge.

It is time now to turn to the non-ferrous alloys, such as brass, phosphor bronze, and beryllium bronze. They are characterized by a face-centred cubic lattice, the crystals of which are intersected at random by the plane of the section. The form of the indentation, which is in general quadratic, is affected by the crystalline structure and may change as the diamond is rotated in relation to the specimen.

Fig. 12 (Plate XXXV) represents phosphor bronze (composition: tin

4·5–6, phosphorus 0·02–0·40, lead 0·02% maximum, copper remainder), and a number of impressions have been made on the same crystal but with varying orientation as the specimen was turned 15° in relation to the diamond after each setting. The load was 20 g., and the hard-

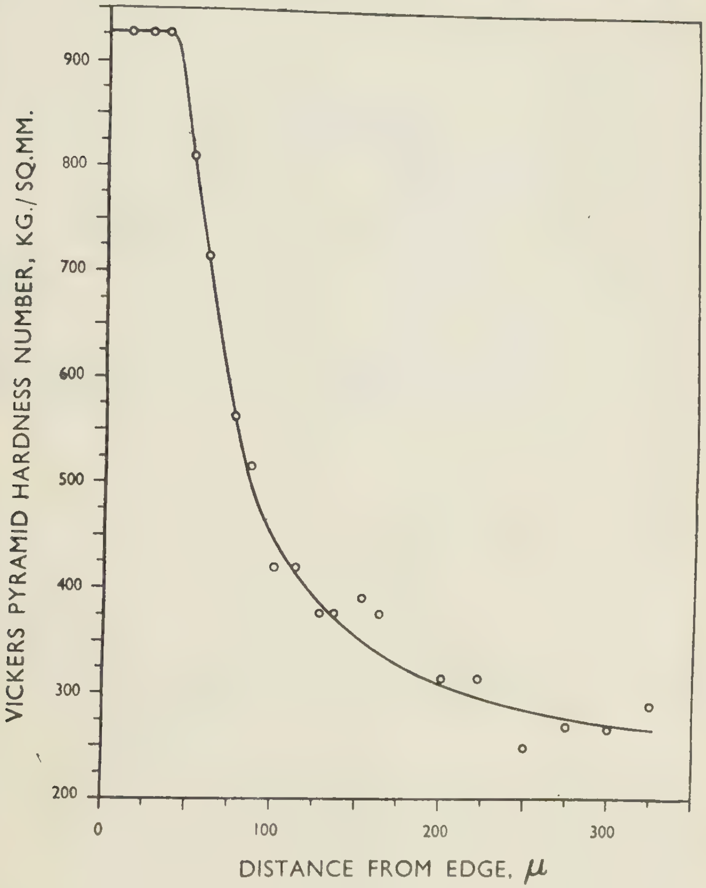

Fig. 11.—Nitrided Steel. Load 50 g.

ness figures for the light and dark areas are 129 and 183. It will be noted that the impressions may have concave or convex sides, and that the light-coloured area is not of uniform hardness. Fig. 13 (Plate XXXV) is of another sample, and here again a difference was noted between the hardness of the light and dark areas. The load was 20 g., and the hardness figures are 140 and 170, respectively. Fig. 14 (Plate XXXV)

is of beryllium bronze (composition: beryllium 2·0, cobalt 0·25%, copper remainder), which forms an α solid solution at high temperatures which is retained on quenching; at this stage the hardness figure was 149 with a load of 10 g. When heated to 200° C. or over, the α crystals darken in a lattice-like pattern, as shown in Fig. 15 (Plate XXXV). Here the load was 10 g. and the hardness figure is 340.

Fig. 16 (Plate XXXVI) is of hard-rolled brass, and the average hardness figure is 117 with a load of 10 g. Some irregularities can be seen in the shape of the impressions, and this is probably due to the orientation of the crystalline axes. The conspicuous twins may be noted.

Finally, it may be interest to describe the determination of the hardness of thin sections of metal which do not lend themselves to the more usual methods. Fig. 17 (Plate XXXVI) shows a section of two safety-razor blades. On one, a number of indentations were made with loads of 5, 10, 20, 50, 100, 200, and 500 g., and on the other a chain of impressions was formed under a load of 20 g. in order to check the uniformity of the hardness throughout the specimen. This was found to be constant, with a hardness figure of 742. Fig. 18 (Plate XXXVII) is a section through nickel plate on brass, and the hardness figure for the nickel was found to be 515 under a load of 50 g. Fig. 19 (Plate XXXVII) shows a piece of tin-foil about 50 μ thick, sandwiched between pieces of brass and gun-metal. The load used was 5 g. and the hardness figure for the tin was found to be 23.

Acknowledgements.

The experiments were carried out in the laboratory of Cooke, Troughton, and Simms, Limited, at York, with the helpful assistance of Messrs. J. W. Harrison and W. S. Hall.

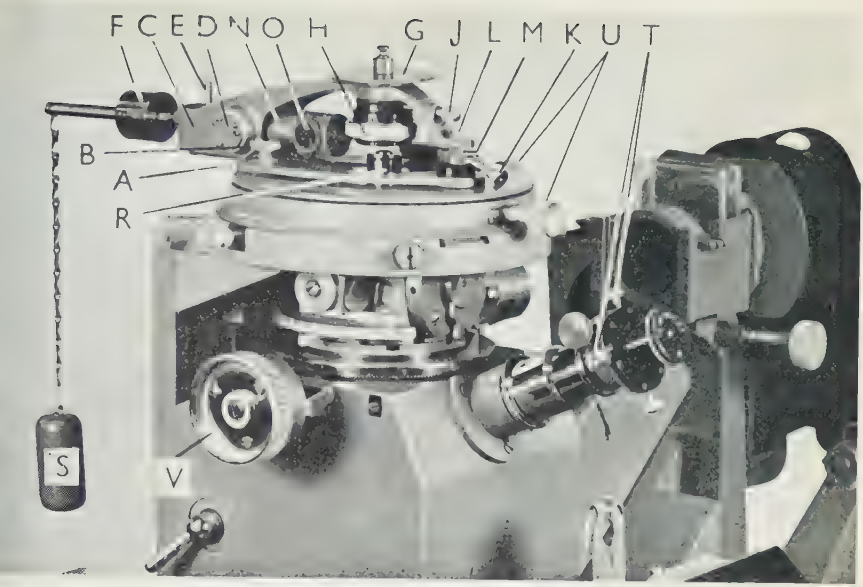

FIG. 1.—The Micro-Hardness Tester as Fitted to a Standard Vickers Projection Microscope.

FIG. 2.—Testing Unit and Accessories in Stowage Box, and Set of Weights.

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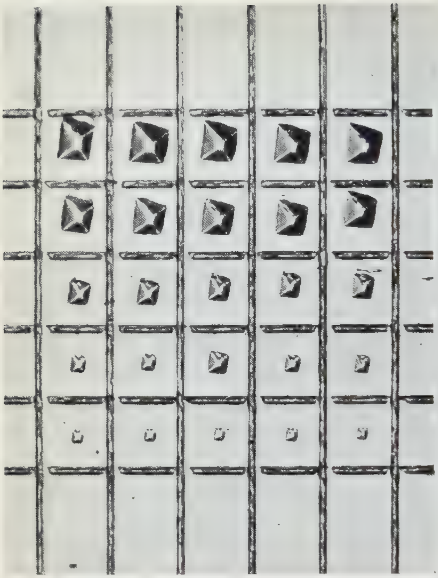


FIG. 3.—Squares of 0·02 mm. Sides Ruled on Austenitic Stainless Steel to Indicate Precision with which the Impressions can be Made. The loads were 50, 40, 20, 10, and 5 g. ×475.

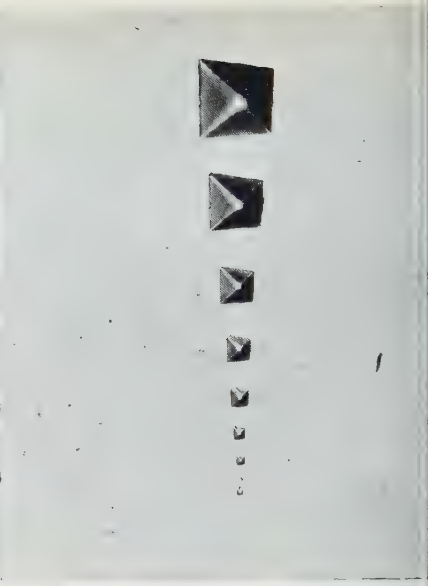


FIG. 4.—Austenitic Stainless Steel. Loads: 200, 100, 50, 20, 10, 5, 2, and 1 g. ×475.

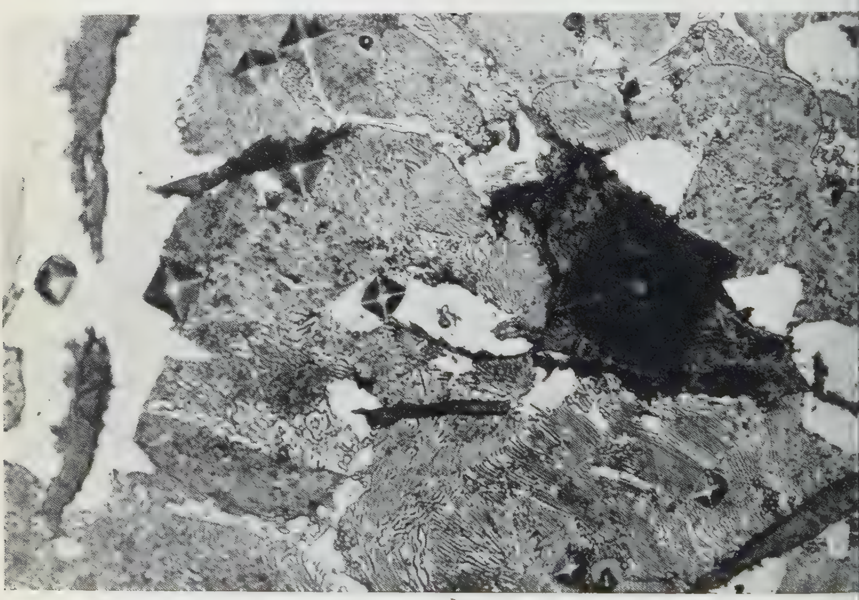
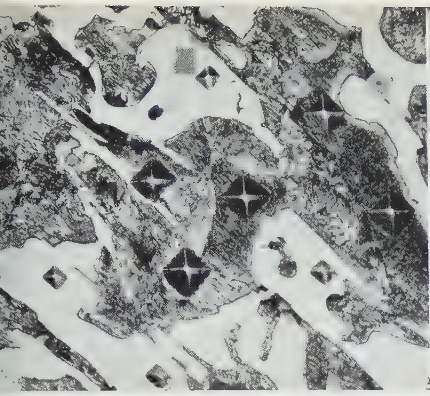


FIG. 6.—Grey Cast Iron. Load 20 g. ×475.



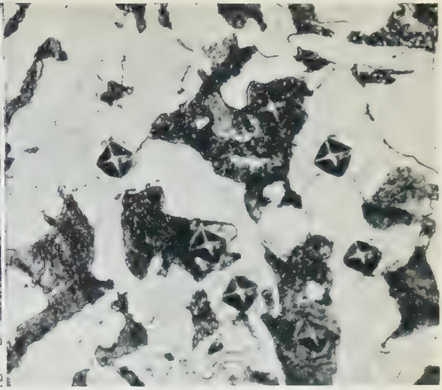
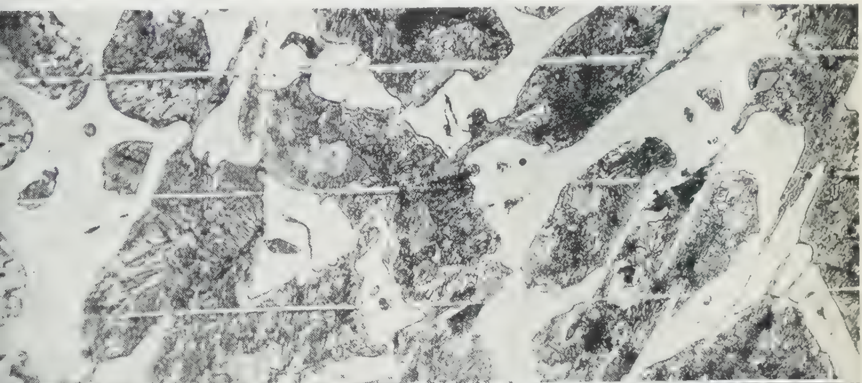


FIG. 7.—Chilled Cast Iron. Load 50 g. ×350.

FIG. 8.—Vickers "A" Steel. Load 10 g. ×475.



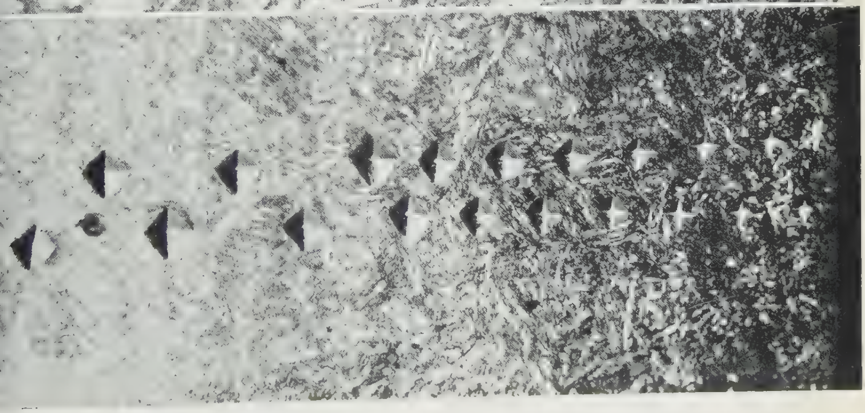


FIG. 9.—Chilled Cast Iron. Scratch tests at loads of 2, 1, and 0·5 g. ×350.
FIG. 10.—Nitralloy Steel, Chemically Hardened from Edge. Load 50 g. ×350.

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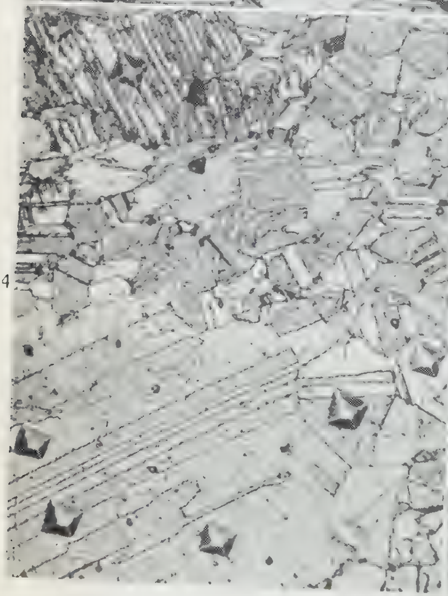
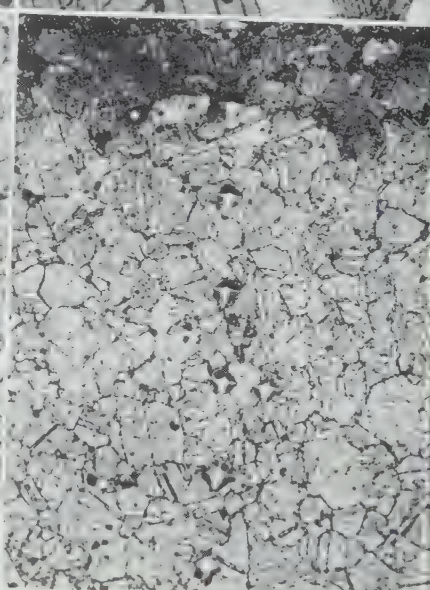
14  

FIG. 12.—Phosphor Bronze. Diamond rotated 15° at each setting. Load 20 g. ×350.
FIG. 13.—Phosphor Bronze. Load 20 g. ×350.
FIG. 14.—Beryllium Bronze, Before Final Heat-Treatment. Load 10 g. ×475.
FIG. 15.—Beryllium Bronze, After Final Heat-Treatment. Load 10 g. ×475.

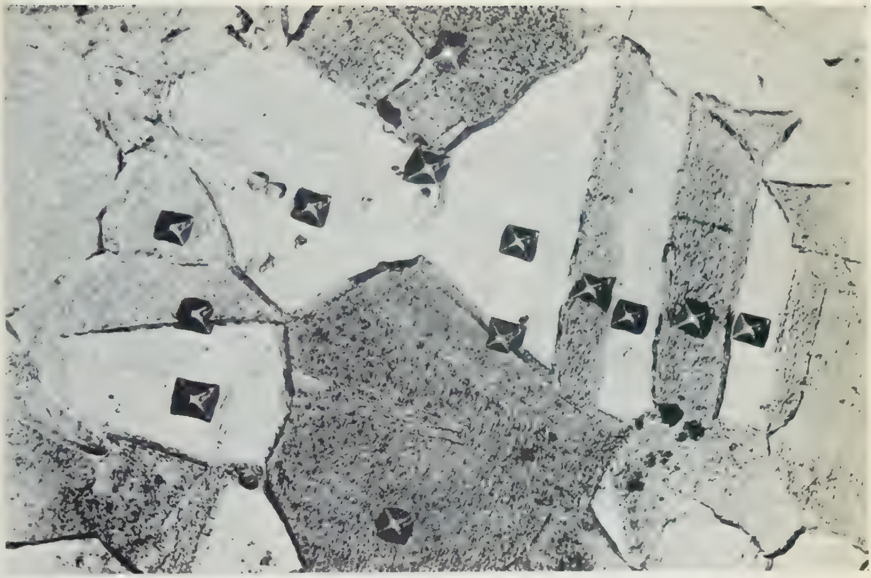


FIG. 16.—Hard-Rolled Brass. Load 10 g. ×475.

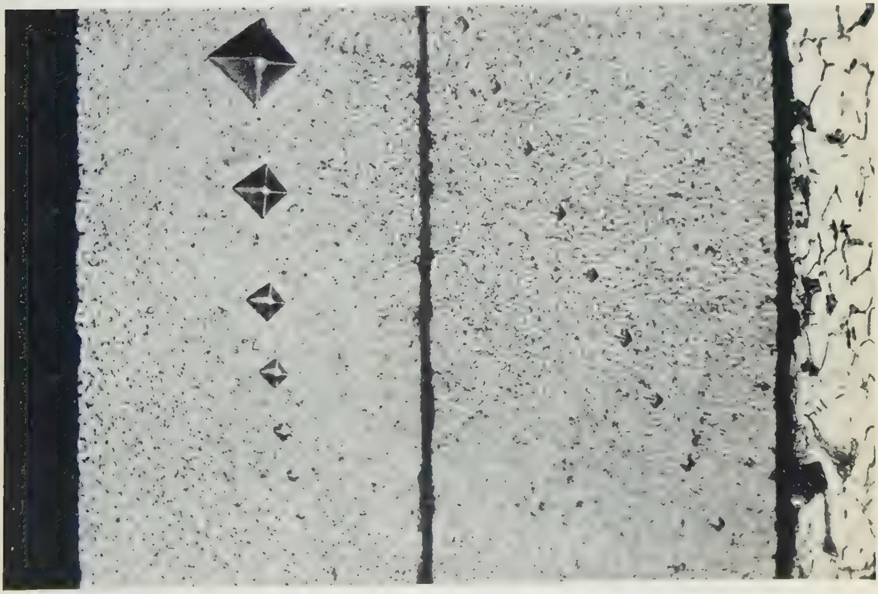


FIG. 17.—Two Safety-Razor Blades. Variable loads on left, 20 g. on right. ×350.

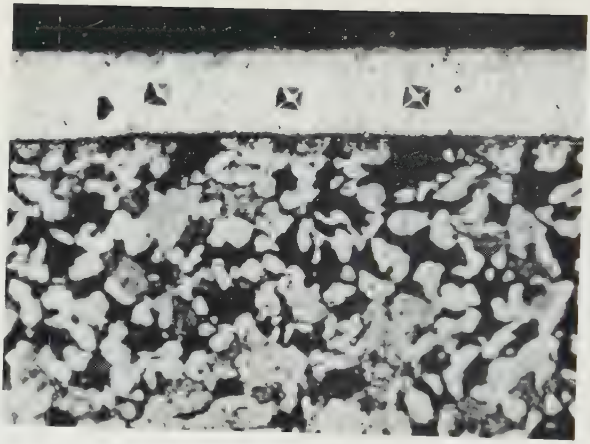


FIG. 18.—Nickel Plating on Brass. Load 50 g. ×350.

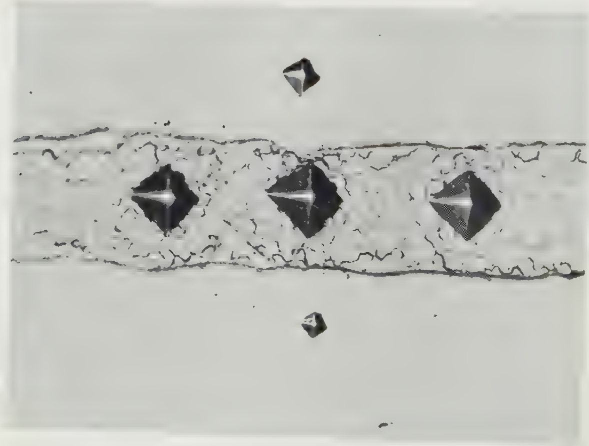


FIG. 19.—Tin-Foil, 0·05 mm. thick, Between Plates of Brass and Gun-Metal. Load 5 g. ×475.

THE EFFECT OF CRYSTAL ARRANGEMENT ON "SECONDARY RECRYSTALLIZATION" IN METALS.* 1132

By J. S. BOWLES,† M.Sc., and W. BOAS,‡ D.Ing.

SYNOPSIS.

The effects on secondary recrystallization of the amount of deformation and of the mode of arrangement of the crystals have been separated by studying this phenomenon in specimens which had been given the same amount of deformation by two types of rolling, namely straight rolling and compression rolling. Straight-rolled specimens possessed the usual type of preferred orientation, while the compression-rolled specimens possessed fibre structures.

The experiments were carried out with copper, silver, and two grades of aluminium.

It was found that secondary recrystallization can occur regardless of the mode of arrangement of the crystals, but that it occurs more rapidly the greater the similarity between their orientations. Reasons are advanced for the conclusion that this is caused by a difference in the ease with which the secondary crystals can grow rather than in their tendency to grow.

It was also found that neither recrystallization nor secondary recrystallization alter the mode of arrangement of the crystals existing after deformation. It is shown that the existence of such an orientation relationship is not necessarily evidence in favour of Burgers's "nuclear spot" theory.

I.—INTRODUCTION.

THE occurrence of grain growth in metals after recrystallization is a phenomenon which has received considerable attention in the past, but nevertheless a satisfactory explanation which accounts for all the observed characteristics of the phenomenon has not yet been found.

A very comprehensive review of the literature on this subject has been made by W. G. Burgers.[1] Burgers makes a distinction between two different modes of grain coarsening. On heating after recrystallization, growth of many grains may occur, and the structure then becomes uniformly coarser. The term "grain growth" as used by Burgers, refers to this phenomenon. However, after very severe deformation only a small number of grains grow, and these attain a very large size, whereas the remaining crystals seem to be unchanged in size. Burgers calls this second type of growth "secondary recrystallization".

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† School of Metallurgy, University of Melbourne, Australia.
‡ Formerly School of Metallurgy, University of Melbourne; now Section of Tribophysics, Council for Scientific and Industrial Research, Melbourne, Australia.

There are, of course, cases in which both types of growth occur simultaneously. It is not possible at present to decide whether there is any fundamental difference between these two types of growth, but we think that it is advisable to make the distinction until the matter can be decided. These two terms will therefore be used in the sense just described.

Considering secondary recrystallization, there is fairly general agreement that the following two observations have been established [2]:

(1) The orientations of the secondary crystals produced in a material with a pronounced primary recrystallization structure are very similar to each other, but differ from the orientation of the primary recrystallization structure.

(2) The presence of a pronounced primary recrystallization structure * seems to favour the occurrence of secondary recrystallization.

Burgers and Basart [3] and Burgers [1] have shown that the secondary crystals which grow in rolled aluminium single crystals (75% reduction) all have a similar orientation, and that this orientation is closer to that of the deformation structure than to that of the primary recrystallization structure.

In polycrystalline iron–nickel alloys [4, 6, 7, 8, 9, 10] the orientation of the secondary crystals is obtained from the primary structure (100) [001], by a rotation of approximately 30° about the rolling direction. If the amount of rotation were exactly 30°, this orientation would be described by (120) [001].

With copper the position is not as clear. Dahl and Pawlek [4] found the same orientation of the secondary crystals as in iron–nickel alloys, whereas Cook and Richards [5] claim that it is (110) [112], i.e. identical with that of rolled copper, but different from that of recrystallized copper (100) [001].

There would seem to be no doubt, therefore, of the validity of observation (1) above. This observation has led some workers to believe that secondary recrystallization occurs by the production of new nuclei and is thus, in this respect at least, a different phenomenon from grain growth.

Regarding observation (2) above, namely that the presence of a pronounced primary recrystallization structure favours the occurrence of secondary recrystallization, it is our opinion that this is not necessarily justified by the experimental results. This conclusion is based on the observation that whenever secondary recrystallization occurs, a preferred orientation is present. However, this cannot be taken to imply that

* Recrystallization in its usual sense is called " primary recrystallization " to avoid confusion with secondary recrystallization.[1]

secondary recrystallization would not occur if the preferred orientation were not present. According to the present theories, it is the amount of deformation prior to primary recrystallization which is the controlling factor in secondary recrystallization. The existence of a preferred orientation may only indicate a large amount of deformation.

It has been the object of this investigation to separate as far as possible the effects of amount of deformation and of mode of arrangement of crystals and so to determine whether a pronounced recrystallization structure really favours the occurrence of secondary recrystallization.

II.—Experimental.

Two sets of specimens were prepared for each metal used, having the same amount of deformation, but differing with respect to the arrangement of the crystals. In one set all the crystals had nearly the same orientation, while in the other set there was only rotational symmetry about a certain direction, i.e. a fibre structure was present. The former set was prepared by straight rolling, the latter by "compression rolling".[11] * On allowing primary recrystallization to occur in these specimens, the modes of arrangement of the crystals were not altered, i.e. a fibre structure remained a fibre structure, and a straight-rolling structure remained one in which two directions were nearly parallel in all the crystals.† It was thus possible by further annealing to study the effect of the different modes of arrangement of the crystals on secondary recrystallization. These experiments were repeated a sufficient number of times to ensure reproducibility.

1. *Preliminary Treatment of Materials.*

The metals used for these experiments were copper, silver, and two grades of aluminium.

The only preliminary treatment given to the copper bar, which was of commercial purity (about 99·95%), was an anneal at 700° C. for ½ hr.

Silver sheet, which was 999 fine, was obtained from the Royal Mint (Melbourne). Since it has been shown that small concentrations of oxygen have a marked retarding effect on grain growth in silver,[12] the sheet was re-cast *in vacuo*. The cast ingot consisted of very large grains. To eliminate the casting structure, the ingot was subjected to alternate compression and annealing (½ hr. at 400° C.) *in vacuo*. This treatment refined the grain-size to about 0·1 mm.

* In compression rolling, the rolling direction is random. Each pass through the rolls is made in a different direction. It is therefore a different process from cross rolling.

† The crystallographic nature of these directions and of the fibre axis may, of course, change on recrystallization.

The specimens of " pure " aluminium (99·98%, with a trace of iron) were cut from a large ingot and then subjected to alternate compression and annealing ($\frac{1}{2}$ hr. at 400° C.).

The commercial aluminium (iron 0·5, silicon 0·2, copper 0·07%) was available in the form of annealed sheet $\frac{1}{8}$ in. thick. Some experiments were also carried out on single crystals, which were prepared from this material by the method of annealing after critical straining.

2. *Rolling.*

After the preliminary treatment, the specimens were cut in half to produce two identical pieces for rolling. One half was straight rolled, care being taken during rolling to keep the same side up and the same end to the front; the other half was compression rolled. Both specimens were rolled at the same time so that they received exactly the same passes.

The thickness of the specimens prior to rolling was $\frac{1}{2}$ in. in the case of copper, silver, and " pure " aluminium, and $\frac{1}{8}$ in. in the case of commercial aluminium. The reduction in thickness by rolling was 97% for copper, 96% for silver, 95% for " pure " aluminium, and 85% for commercial aluminium.

In the experiments on commercial aluminium, the single crystals and the polycrystalline materials were rolled simultaneously.

3. *Annealing.*

Individual specimens were used for all the treatments, i.e. the same specimen was not given a short-time anneal and then reheated for a further period. Moreover, in all experiments the straight- and compression-rolled specimens were annealed together, so that there was no possibility of a difference in treatment of the two types of specimen.

In the case of the copper and silver, all annealing was carried out in evacuated silica tubes. These tubes were not sealed off until the specimens had been at the annealing temperature for at least $\frac{1}{2}$ hr. The pressure in the tubes at the time of sealing was not greater than 1×10^{-3} mm. mercury. The surface of the copper specimens was always bright and free from scale, even after annealing at 700° C. for 3 weeks. It was not necessary to take any precautions to avoid oxidation of the aluminium specimens.

4. *Examination of Annealed Specimens.*

The annealed specimens were then examined in order to determine (1) the extent to which secondary recrystallization had occurred in the

various specimens, and (2) the shape of the secondary crystals and their orientation relative to the specimen.

In general, the extent of secondary recrystallization and the shapes of the secondary crystals could be determined by visual examination of the macrostructure. However, in certain cases it was desired to have more detailed information, and for this purpose the specimens were polished electrolytically and etched (see Table I).

Two methods were used for determining the orientations of the secondary crystals. In general, the orientations were determined from Laue back-reflection photographs. However, in certain cases it was found to be more convenient to determine the orientations by the etch-pit method.[13] The etchants used to develop the etch pits, and the crystallographic nature of these pits, are shown in Table I.

Table I.

Metal.	Electrolytic Polishing Method Used.	Etchant Used for Developing Crystallographic Etch Pits.	Crystallographic Planes Developed.
Copper	P. Jacquet [14]	1 part HCl, 1 part H_2O, mixture saturated with $FeCl_3$.	(100)
Silver	L. Gilbertson and O. Fortner [15]	5 g. ammonium persulphate, 50 ml. 880 ammonia, 100 ml. H_2O.	(100) and (111)
Aluminium	A. de Sy and H. Haemers [16]	45 ml. HCl, 15 ml. H_2F_2, 15 ml. HNO_3, 25 ml. H_2O.	(100)

The chief advantages of this method are its rapidity, and its application to crystals which are too small to allow adjustment in an X-ray beam. The accuracy is about the same as that obtainable with the Laue method, namely, about $\pm 1°$. It does not give accurate results, however, if the orientation of the crystal is such that reflections occur at small angles to the plane of the surface of the specimen.

III.—Results.

It was found in all cases that the straight-rolled specimens had a slightly higher hardness than the corresponding compression-rolled specimens. The results of the hardness determinations will be discussed in the Appendix.

1. *Arrangement of Crystals After Rolling and Primary Recrystallization.*

X-ray transmission diagrams were made from all specimens in the rolled condition and after recrystallization. The grain-size of the

recrystallized aluminium and silver was so large that it was necessary to move the specimens during exposure to allow more crystals to be irradiated.

Complete pole figures were constructed only for the compression-rolled polycrystalline specimens. They all possessed fibre structures, and these will be described elsewhere. It was not considered necessary to make complete determinations of the structures of the straight-rolled specimens since these are always of the type where two directions are nearly parallel throughout the specimen.[1, 3]

These modes of arrangement were not altered by primary recrystallization.

In the single crystals of commercial aluminium, compression rolling did not produce a fibre structure but only distorted the original orientation to about the same extent as straight rolling. The mode of arrangement in these specimens was actually very similar to that present in the straight-rolled single crystals. Some typical X-ray transmission diagrams of these specimens are reproduced in Figs. 1–6 (Plate XXXVIII).

The attempts to produce the desired arrangements in recrystallized "pure" aluminium were not successful. Although the preferred orientations were present in the rolled condition, after primary recrystallization neither the straight-rolled nor the compression-rolled specimens showed marked preferred orientation of either type, and there was in fact very little difference between the two sets of specimens.

2. *Rate of Secondary Recrystallization.*

(a) *Copper.*

Table II shows the results of annealing experiments carried out at 700°, 850°, and 1030° C. for various periods. The behaviour of a

TABLE II.—*Secondary Recrystallization in Rolled Copper* (97% *reduction*).

Temperature, ° C.	Time.	Straight Rolled.	Compression Rolled.
700	½ hr.	Prim. recryst. only	Prim. recryst. only
	1 day	Grain growth	Grain growth
	1 week	Second. recryst. complete	Further grain growth
	3 weeks	,,	,,
850	3 days	Second. recryst. started	Grain growth
	1 week	Second. recryst. complete	,,
1030	1 hr.	Second. recryst. complete	Grain growth

typical series of specimens heated at 700° C. is shown in Figs. 7–12 (Plate XXXIX).

Since secondary recrystallization did not occur in any of these compression-rolled specimens, further experiments were carried out on specimens which had been more severely deformed (98·6% reduction). Secondary recrystallization did not occur in any of these specimens after heating for 1 hr. at 1030° C.

(b) *Silver.*

Two series of experiments were carried out. In the first series, the specimens were heated at 550° C. for periods of ½ hr., 1 day, 1 week, and 3 weeks. The grain-size of both types of specimen increased, but there was no difference in their behaviour. Secondary recrystallization had commenced in both types of specimen after 1 week, but even after 3 weeks the secondary crystals were not very large, and it seems that very long periods at 550° C. would have been needed to complete secondary recrystallization.

The results of the second series of experiments, in which the specimens were heated at 700° C., are shown in Table III. The behaviour of a typical series of these specimens is shown in Fig. 13 (Plate XL).

Table III.—*Secondary Recrystallization in Rolled Silver* (96% *reduction*).

Temperature, ° C.	Time.	Straight Rolled.	Compression Rolled.
700	½ hr.	Prim. recryst. only	Prim. recryst. only
	1 day	Second. recryst. complete (several large crystals)	Grain growth only
	3 days	,,	Second. recryst. complete (smaller second. crystals)

(c) "*Pure*" *Aluminium.*

Numerous experiments were carried out with the "pure" aluminium at 350°, 450°, and 600° C. for times ranging from 5 min. to 6 weeks. In none of these experiments was the behaviour of the two types of specimen different, and in no case were large secondary crystals produced, but a gradual increase in grain-size was always observed.

(d) *Commercial Aluminium.*

All the rolled specimens were heated at 600° C., for periods of 5 min., ½ hr., and 1 hr. The changes which occurred in the specimens can be seen from Fig. 14 (Plate XL), which is a photomacrograph of a typical series.

After 5 min., only primary recrystallization had occurred, and the grain-size was uniformly small in all the specimens.

After $\frac{1}{2}$ hr., secondary recrystallization had begun in the rolled single crystals and in the compression-rolled polycrystalline specimens, and it was just observable in the straight-rolled polycrystalline specimens.

After 1 hr., secondary recrystallization was complete in all single crystals, but was not so far advanced in the polycrystalline specimens. The extent of the secondary recrystallization, as judged by the proportion of primary structure consumed, was greater in the compression-rolled than in the straight-rolled polycrystalline specimens.

(e) *Summary of Observations on the Rate of Secondary Recrystallization.*

The above-quoted results appear to justify the following generalizations :

(1) Secondary recrystallization can occur in both types of specimen, i.e. regardless of the mode of arrangement of the crystals in the primary structure.

(2) The rate of secondary recrystallization is greater in specimens in which the crystallographic axes of the crystals are nearly parallel than in specimens in which the preferred orientation is less pronounced or of a lower symmetry, e.g. a fibre structure.

That secondary recrystallization was not observed in compression-rolled copper may possibly be due to the fact that the rate was too small to be detected in our experiments.

If this explanation can be accepted, the second generalization is supported by the behaviour of copper and silver. It is also supported by the contrast in behaviour of single-crystalline and polycrystalline commercial aluminium, since the preferred orientations are very much more pronounced in the former specimens. Straight- and compression-rolled single crystals have the same type of preferred orientation and the same rate of secondary recrystallization.

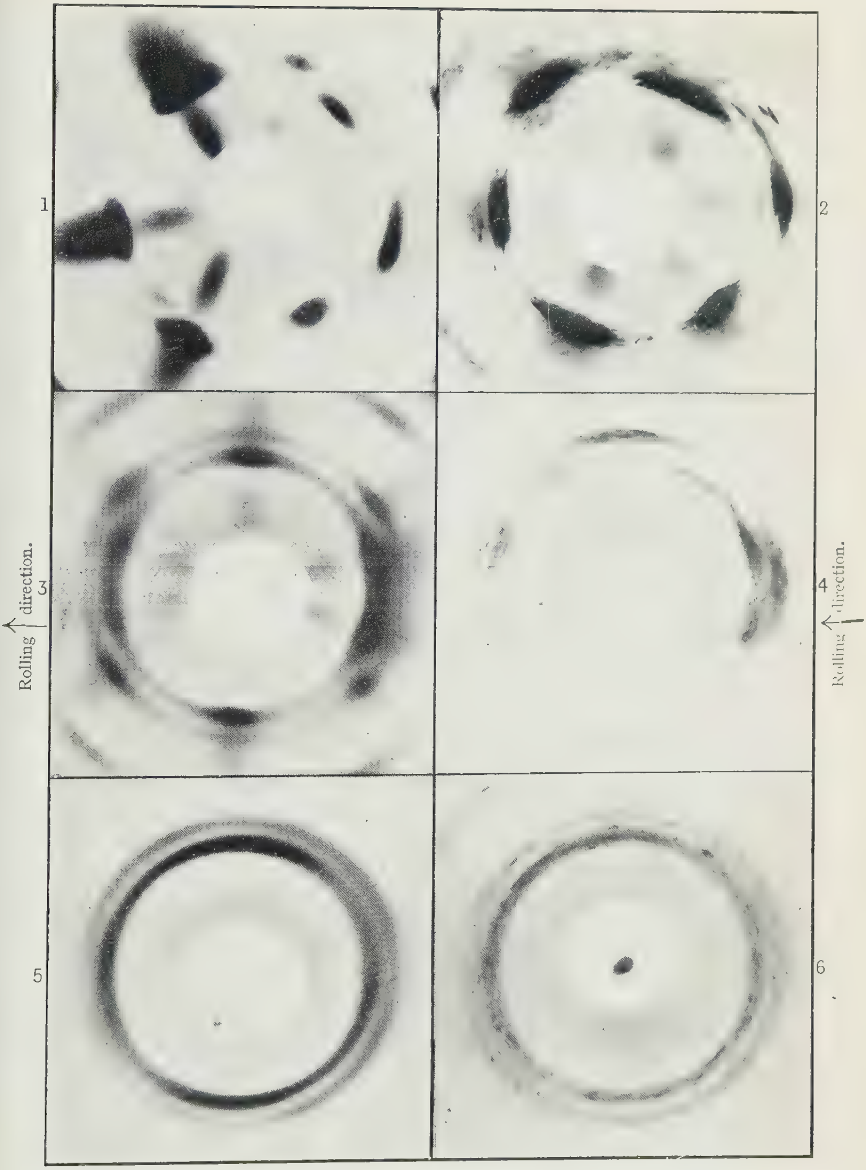
The behaviour of the polycrystalline commercial aluminium is, however, anomalous, but the preferred orientation was not very pronounced in the straight-rolled specimens.

No evidence is obtainable from the experiments with "pure" aluminium, since no marked preferred orientation was present in any of the recrystallized specimens.

3. *Shape of Secondary Crystals During Growth.*

It was noticed that in some cases the growing secondary crystals assumed characteristic shapes.

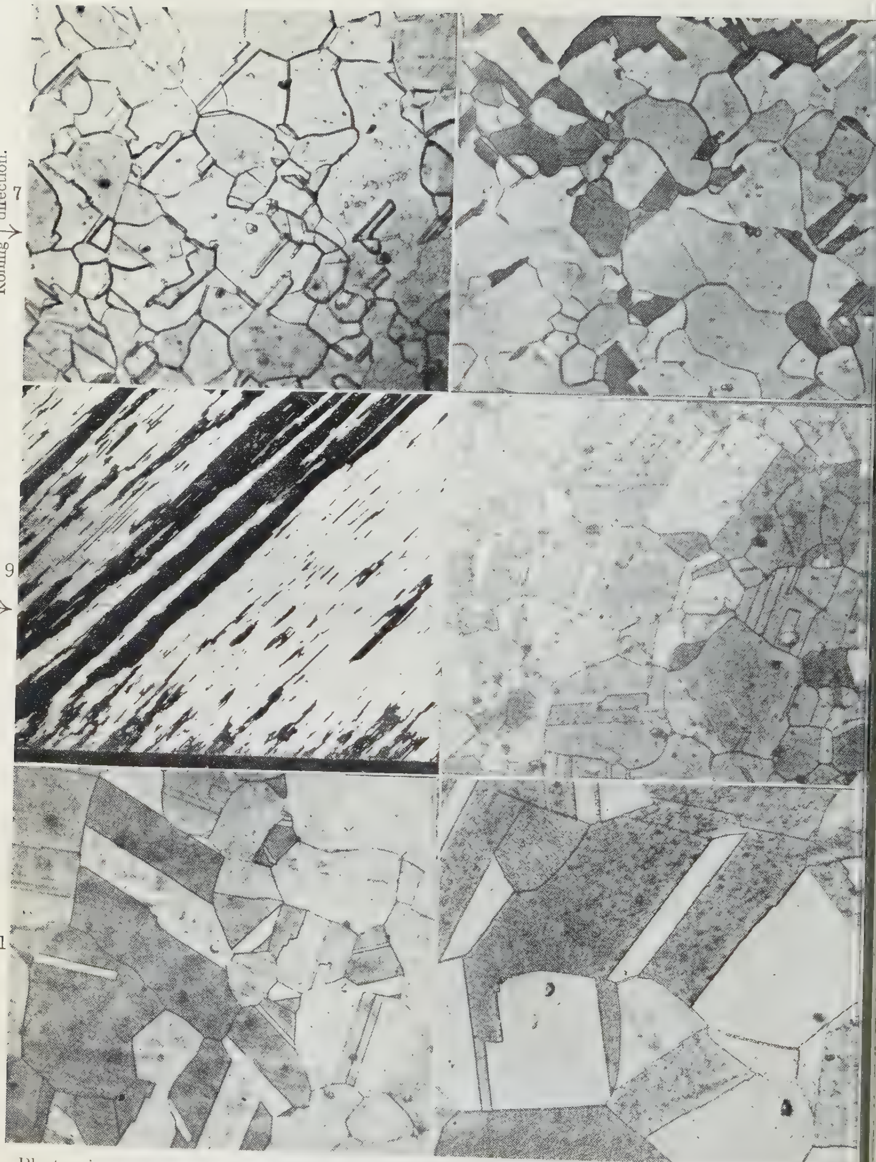
Typical examples of the shapes observed in straight-rolled copper are shown in Figs. 15 and 16 (Plate XLI). It is apparent from these photographs that the secondary crystals do not grow at the same rate



Typical X-Ray Diagrams of Commercial Aluminium. Ag $K\alpha$ radiation.

Fig. 1.—Compression-rolled single crystal.
Fig. 2.—Same after primary recrystallization at 600° C.
Fig. 3.—Straight-rolled polycrystalline material.
Fig. 4.—Same after primary recrystallization at 600° C.
Fig. 5.—Compression-rolled polycrystalline material.
Fig. 6.—Same after primary recrystallization at 600° C.

[*To face p.* 508.

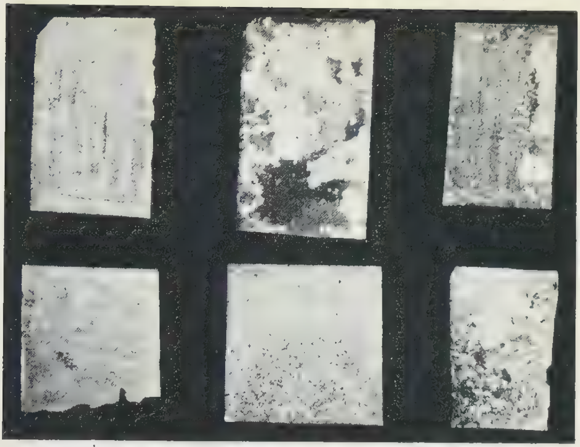


Photomicrographs of Copper Specimens Annealed at 700° C. Fig. 9 × $2\frac{1}{2}$, others all × 250.

Figs. 7–9.—Straight-rolled specimens annealed for $\frac{1}{2}$ hr., 1 day, and 1 week, respectively. Not that only one type of twin is evident in the secondary crystal shown in Fig. 9.

Figs. 10–12.—Compression-rolled specimens annealed for $\frac{1}{2}$ hr., 1 day, and 1 week, respectively.

a b c



d e f

Fig. 13.—Photomacrographs of Silver Specimens Annealed at 700° C. Approx. natural size.

(a)–(c) Straight-rolled specimens annealed for ½ hr., 1 day, and 3 days, respectively.

(d)–(f) Compression-rolled specimens annealed for ½ hr., 1 day, and 3 days, respectively.

a b c d

Time of Annealing.

5 min.

½ hr.

1 hr.

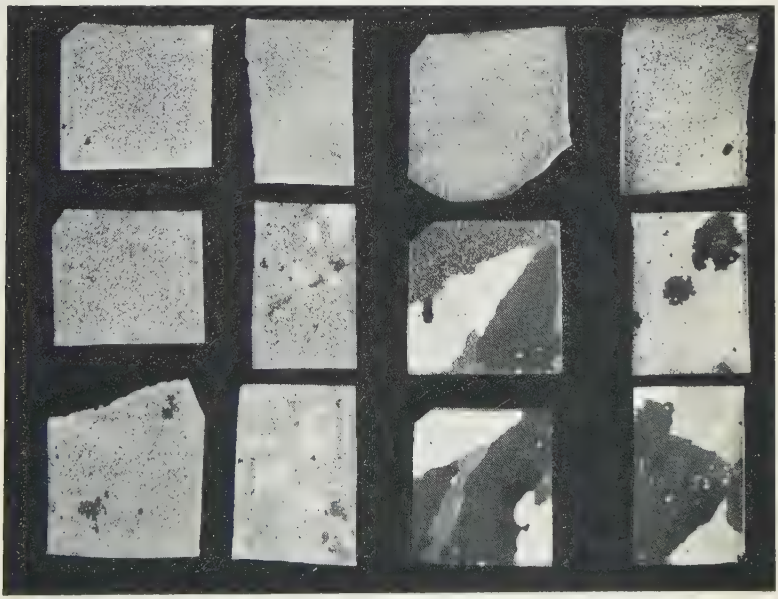


Fig. 14.—Photomacrographs of Commercial Aluminium Specimens Annealed at 600° C. × ⅔.

(a) Straight-rolled polycrystalline material.
(b) Compression-rolled polycrystalline material.
(c) Straight-rolled single crystals.
(d) Compression-rolled single crystals.

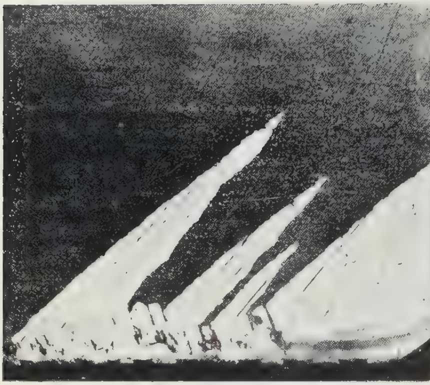
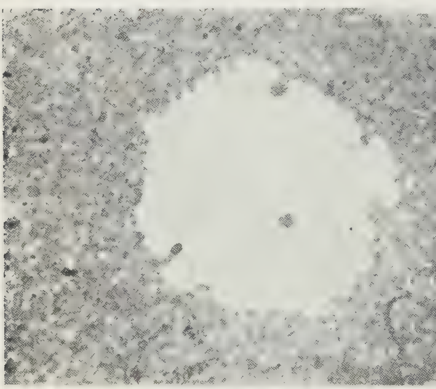


FIG. 15.

FIG. 16.

Photomacrographs of Straight-Rolled Copper Specimens, Showing Typical Shapes Assumed by Secondary Crystals During Growth. × 2.



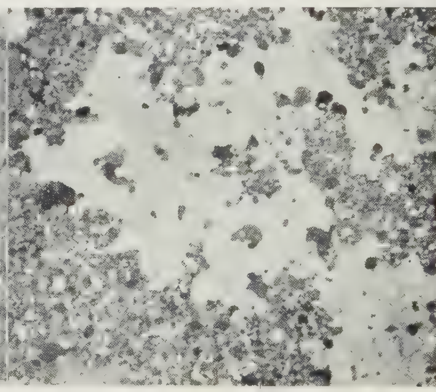


FIG. 17.—Straight rolled.

FIG. 18.—Compression rolled.

Photomacrographs of Commercial Aluminium Specimens, Showing the Shapes Assumed by Secondary Crystals During Growth. × 8.

in all directions, and the shape of a crystal at any stage in its growth will be determined by this anisotropy. It is not possible without knowing the position of the nucleus to tell from the shape of a crystal in which direction growth has taken place most rapidly. However, it is likely that this direction, in the triangular crystals of Fig. 15, lies in between the sides of these triangles. It will be noticed that the intersection of the twin plane with the surface also lies in between these sides, and that this is evidently close to, if not identical with, the direction of maximum rate of growth. In this case, Cook and Richards's statement [5] that the rate of growth is greatest in the (111) plane of the secondary crystals would be confirmed. This is also evident from Fig. 16, which shows moreover that the secondary crystal and its twin develop simultaneously and at the same rate.

In aluminium, the shape of the secondary crystals varied markedly with the mode of arrangement of the primary crystals (Figs. 17 and 18, Plate XLI). In all the specimens in which the crystals were similarly orientated, the secondary crystals grew roughly in the form of spheres, advancing into the primary structure along a more or less sharply defined interface. In the compression-rolled polycrystalline specimens, however, the secondary crystals had very irregular boundaries, and many small crystals of the primary structure remained embedded in them.

4. *Orientations of Secondary Crystals.*

(a) *Copper.*

Fifteen determinations were made on 11 secondary crystals. It was found that the crystals had one or other of two different orientations, and both orientations were found in different parts of some of the crystals. There was a twin relationship between the two orientations. It is not possible to decide which of the two orientations in twin positions is the parent orientation.

Fig. 19 is a pole figure of the orientations. In this Figure one group of orientations has been twinned in the appropriate manner to match the other group.

The range of scatter of these results makes it impossible to describe them adequately by means of a single set of Miller indices, although previous workers [4, 5] have done so without giving any quantitative indication of the ranges of scatter of their results. It is therefore not possible to make a proper comparison between the reported orientations and those found in the present investigation.

The orientation (120) [001] reported by Dahl and Pawlek [4] has been plotted into Fig. 19, and although there is some similarity, the agreement

is not good. The orientation (110) [112] reported by Cook and Richards [5] is also shown, but in this case there is no similarity at all. This orientation lies closer to our twin orientation, but the discrepancy is still greater than that between (120) [001] and our results. It is therefore clear that neither of the reported orientations is suitable to describe those observed in this work.

A definite relationship has been found to exist between the orientation of the secondary crystals and the primary structure. This relation is

Fig. 19.—Stereographic Projection of Orientations of Secondary Crystals in Straight-Rolled Copper.

Dahl and Pawlek's orientation.
Cook and Richards's orientation.

suggested by Figs. 7 and 15 and 16 (Plates XXXIX and XLI), which show that the line of intersection between the twinning plane and the rolling plane is at 45° to the rolling direction in both the primary structure and the secondary crystals. This can only be the case if the two orientations are related in one of two ways. The twin plane in the secondary crystal is produced from that of the primary structure either by a rotation about the visible line of intersection with the surface of the specimen, or by a rotation about its normal, i.e. in its own plane. This latter relation proved to be the correct one. The projections of the

two secondary orientations showed that the {111} plane common to both coincided with one of the {111} planes of the primary structure. Thus, the observed secondary orientations can be produced from the primary structure, (100) [001], by a rotation about a [111] direction. This is illustrated in Fig. 20.

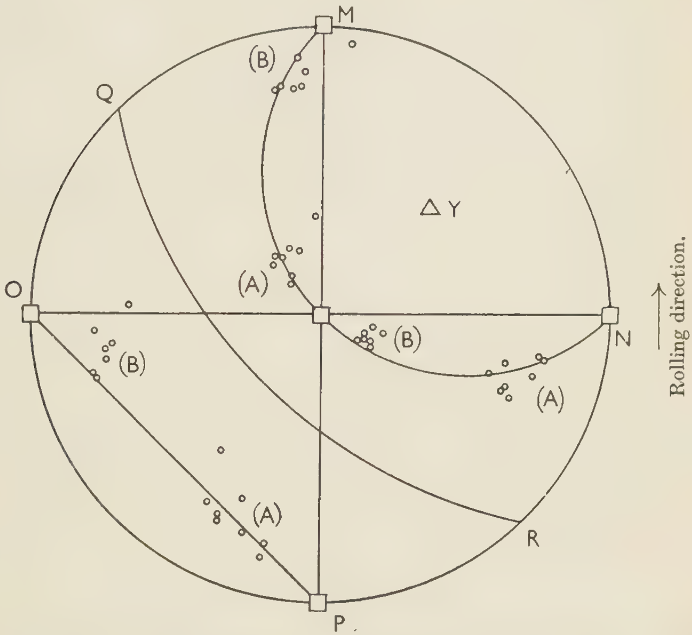


Fig. 20.—Generation of the Two Secondary Orientations "A" and "B" Observed in Copper from the Cube Structures (001) [100] by Rotation about the [111] Direction *Y*.

The curves *MN* and *OP* are at 55° to *Y*. The great circle *QR*, at 90° to *Y*, is the plane in which rotation occurs.

(b) *Silver.*

Fig. 21 shows the orientations of 10 of the secondary crystals which grew in 3 straight-rolled specimens. This pole figure shows that all the crystals except one have similar, but not identical, orientations. This one exception proved to be a twin of the others.

The rolling plane in these crystals can be described quite well by the indices (110) since there is little scatter in this plane, as is indicated by the radial scattering of the poles about the circle at 45° from the centre of the projection. The rolling direction cannot be described by Miller indices because of the considerable degree of scattering.

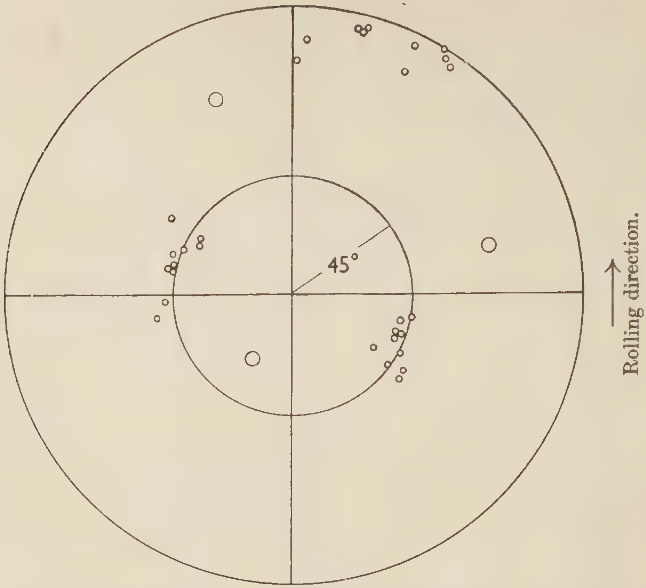


Fig. 21.—Orientations of Secondary Crystals in Straight-Rolled Silver Specimens. The orientation indicated by the large circles is a twin of the main orientation.

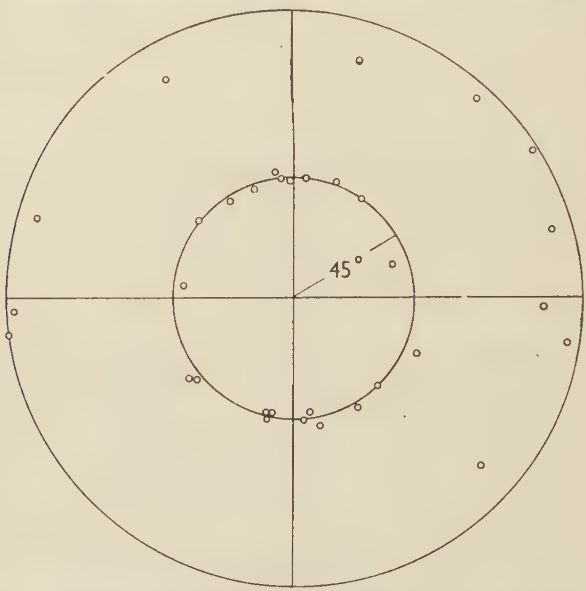


Fig. 22.—Orientations of Secondary Crystals in Compression-Rolled Silver Specimens.

The orientations of the secondary crystals which grew in the compression-rolled specimens are shown in Fig. 22. It can be seen from this pole figure that two of the cubic poles of each crystal lie at about 45° from the centre of the projection and the third lies near the circumference of the projection. This means that in all the secondary crystals the rolling plane is the (110) plane. However, there is no preferred direction in the rolling plane, i.e. a fibre structure is still present after secondary recrystallization.

Fig. 23.—Orientations of Secondary Crystals in Straight-Rolled Polycrystalline Commercial Aluminium.

(c) *Aluminium.*

The orientations of 19 of the secondary crystals which grew in 4 specimens of straight-rolled polycrystalline commercial aluminium were determined by the etch-pit method. Fig. 23 shows that the orientations are similar, which is not in agreement with Dahl and Pawlek's [4] observation that such crystals have a random orientation.

The orientations of the secondary crystals which grew in the compression-rolled polycrystalline specimens are shown in Fig. 24. All the poles lie at approximately 55° to the centre, indicating that a {111} plane lies in the rolling plane. There is, however, no directional preferred orientation within this plane.

The orientations of the secondary crystals which grew in the straight-rolled single crystals were found to be very similar to one another, in agreement with Burgers's observations.[1]

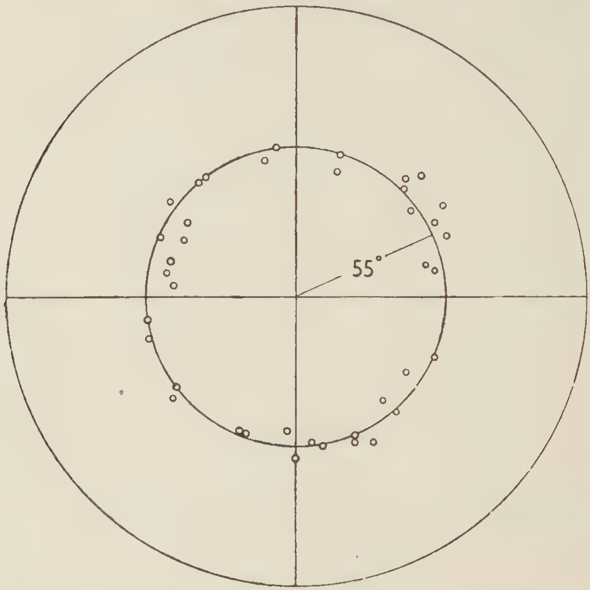


Fig. 24.—Orientations of Secondary Crystals in Compression-Rolled Polycrystalline Commercial Aluminium.

IV.—Discussion of Results.

1. *Effect of Mode of Arrangement of Crystals on Secondary Recrystallization.*

It is now clear that the presence of a preferred orientation can be said to favour the occurrence of secondary recrystallization, and does not simply indicate a large amount of deformation prior to primary recrystallization.

It has been shown that secondary recrystallization occurs more rapidly when the orientations of the crystals of the primary structure are similar. To cause this effect, either the number of growing crystals or the rate of growth must be greater in straight-rolled than in compression-rolled specimens. The number of secondary crystals does not seem to depend on the type of rolling, in fact in the case of silver it was greater in the compression-rolled specimens. The effect

must therefore be caused by a greater rate of growth in the straight-rolled specimens.

This greater rate can be attributed to an increase either in the tendency for the crystals to grow or in the ease with which they can grow.

If the first of these explanations is correct, then the difference in the arrangement of the crystals cannot be the only significant difference between corresponding straight- and compression-rolled specimens after primary recrystallization. There must be an additional difference, possibly in the energy states of the primary crystals. Such a difference could arise as a result of a difference in the state of deformation of the rolled specimens. That this is not the case is shown by the experiments carried out on straight- and compression-rolled single crystals of aluminium (Fig. 14, Plate XL). These had a similar type of preferred orientation, and so any difference in the rate of secondary recrystallization in these specimens would be a direct result of the difference in the type of rolling. No such difference in rate was observed. Furthermore, in all the experiments the grain-sizes of the straight- and compression-rolled specimens after primary recrystallization were the same, indicating again that corresponding straight- and compression-rolled specimens were in the same state of deformation.

A difference in energy states could, however, arise through an effect of the mode of arrangement of the crystals on the redistribution of stresses which occurs during primary recrystallization. This redistribution of stresses [17, 18] is a probable source of the imperfections in primary crystals. From the existing data, it does not seem to be possible to decide whether such a difference in energy states is produced in this way. However, it is our opinion that the difference in the rate of growth is not due to a difference in the tendency for the secondary crystals to grow, but rather to a difference in the ease with which the secondary crystals can grow.

This view is supported by observations which were made on the shapes of growing crystals. In straight-rolled commercial aluminium the secondary crystals apparently grew at a uniform rate in any given direction, while in the compression-rolled specimens the interface between the primary and secondary crystals was very irregular (Figs. 17 and 18, Plate XLI). These differences can be explained if it is assumed that the rate of migration of a boundary in any direction is influenced by the change of orientation which occurs on crossing the boundary in that direction. In the straight-rolled specimens this change of orientation in any direction remains the same throughout the growth, since all the primary crystals are similarly orientated with respect to the specimen. In the compression-rolled specimens, however, the change of orientation

across the boundary of the secondary crystal varies as successive primary crystals are consumed. Apparently, the secondary crystal can only consume easily those crystals which have a favourable orientation, and so growth occurs in an irregular manner, the boundary becoming very irregular and numerous primary crystals not being consumed at all.

Another observation indicating that the relative orientation affects the ease with which secondary crystals can consume primary crystals, is the anisotropy in the rate of growth of the copper secondary crystals. These crystals develop most rapidly along that octahedral plane which has the same orientation in the primary and in the secondary structure except that in the secondary crystals it has been rotated 30° about its own normal. Apparently, the amount of atomic movement required to achieve this rotation is only small, and consequently growth of the secondary crystal in directions lying in this plane can occur quite readily. The octahedral plane, being the plane in which the cohesion is a maximum, can probably undergo rotation without being severely distorted.

As mentioned previously, the two orientations observed in the copper specimens could be produced from that of the primary structure by rotation within a {111} plane. This arises from the fact that the twinning plane and the plane in which rotation of the lattice had to occur were the same. It is not surprising therefore that the two twin orientations could develop at the same rate.

2. *The Role of Orientation Relationships in Burgers's Theory of Secondary Recrystallization.*

It is evident from this investigation that the secondary crystals possess the same type of preferred orientation as the primary structure. Although it is obvious that to be satisfactory any theory of secondary recrystallization must explain this orientation relationship, only that proposed by Burgers and Sandee [2, 19] attempts to do this.

The basic assumption of their theory is that both primary and secondary recrystallization occur by growth of certain regions of the deformation structure, i.e. there is no nucleation in the classical sense of the word, the "nuclei" for both processes being present in the deformed metal.

This assumption implies that: (1) the orientations of both the primary and secondary crystals are present in the deformed specimens; and (2) during primary recrystallization the growing crystals do not consume the whole of the deformed material; those regions of the deformation structure which are subsequently to act as the nuclei for secondary recrystallization remain unconsumed.

The theory explains adequately the persistence of a given mode of arrangement of crystals throughout primary and secondary recrystallization, and also the general observation[1] that the orientation of the secondary crystals is closer to that of the deformation structure than to that of the primary structure.

No definite proof that nucleation does occur in this manner has yet been obtained. However, the authors would like to point out that the essential features of this explanation could be valid, regardless of the mechanism of nucleation for primary recrystallization. The more generally accepted theory of primary recrystallization is that, on heating, strain-free nuclei are formed in the most severely distorted regions of the metal by " true nucleation ". Such a mechanism of primary recrystallization would still be compatible with the suggestion that certain regions of the test-piece are not consumed during the process.

It has been shown that during primary recrystallization a redistribution of the internal stresses occurs, and that the growing primary crystals are therefore put into a state of internal stress or perhaps even plastically deformed.[17, 18] At the same time, stresses in the unrecrystallized matrix are relieved to some extent by this redistribution and also by recovery. Thus, the growing primary crystals, although originally the most perfect, and therefore the most stable parts of the specimen, do not necessarily remain so. When they become less stable than the most perfect parts of the deformed metal, the latter will tend to grow at the expense of the primary crystals, i.e. secondary recrystallization will occur.

This picture of the process of secondary recrystallization does not involve any assumption about the mechanism of formation of primary crystals, and simply suggests, in conformity with Burgers's ideas, that secondary recrystallization occurs by growth of certain parts of the primary structure. The parts in question are those which have been " left over " from the deformation structure. This picture also explains the similarity of orientation between secondary crystals and deformation structures. It will therefore be clear that this similarity of orientation cannot be used as an argument in favour of Burgers's complicated theory of " nuclear spots ".

APPENDIX.

Hardness of Rolled Specimens.

The hardness values of the rolled specimens are shown in Table IV. Each of these values is the mean of at least 10 readings. The standard

deviation of each of the means is also given. It can be seen that the difference in hardness between corresponding straight- and compression-rolled specimens is significant.

Table IV.—*Effect of Straight and Compression Rolling on Hardness.*

Metal.	Reduction by Rolling, %.	Vickers Hardness Number.	
		After Straight Rolling.	After Compression Rolling.
Copper	97	133 ± 0·4	114 ± 1·4
Silver	96	105·7 ± 0·7	100·0 ± 0·6
" Pure " aluminium	95	40·7 ± 0·3	35·9 ± 0·3
Commercial aluminium	0	38·6 ± 0·8	38·6 ± 0·8
	23	48·6 ± 0·7	42·6 ± 0·7
	50	51·5 ± 0·3	45·3 ± 0·5
	74	61·1 ± 2·6	49·7 ± 0·6

It seems to be quite general that the amount of work-hardening is smaller if the direction of deformation is changed than if deformation takes place in one direction only.[20] Brick and Williamson [21] have observed a similar difference in hardness between straight- and cross-rolled copper. They attribute the lower hardness of the cross-rolled specimens to the formation of mechanical twins, which they claim relieves internal distortion. This explanation cannot be accepted, since the only evidence for the existence of mechanical twins in copper given by Brick and Williamson, namely their pole figure data, has been refuted by Barrett.[22] Furthermore, Mathewson and Van Horn [23] were not successful in their attempts to cause twinning of copper crystals by deformation.

A possible explanation for the difference in hardness is that it is due to a difference in the internal stresses set up by the two types of deformation. In this case, it would be expected that the hardnesses of the straight- and compression-rolled specimens would become equal if the internal stresses were relieved by annealing below the recrystallization temperature. The details of these experiments are shown in Table V.

These results show that the hardness and also the difference in hardness of straight- and compression-rolled specimens are reduced by low-temperature annealing. Apparently, the difference in hardness is caused partly by a difference in the internal stresses, but other factors must be taken into account.

TABLE V.—*Low-Temperature Annealing of Commercial Aluminium.*

Treatment.	Vickers Hardness Number.	
	Straight Rolled.	Compression Rolled.
Rolled, 91% reduction	66·6 ± 0·5	57·0 ± 0·6
4 weeks at room temperature	64·3 ± 0·4	54·4 ± 0·3
Same specimens annealed 6 hr. at 150° C.	63·7 ± 0·3	55·3 ± 0·5
Same specimens annealed 3 hr. at 200° C.	56·0 ± 0·1	49·9 ± 0·2

Another possible cause is the difference in the areas of grain boundaries produced by the two types of rolling. It can be shown quite generally that the area of the grain boundaries of formerly equi-axed grains is increased to a greater extent by straight rolling than by cross- or compression-rolling.

This difference in increase of area is small and in itself would not greatly affect the hardness. However, it is not the effect of the grain boundaries during the hardness test which is important, but their effect during deformation. During rolling, the area of boundary surface becomes greater in the straight-rolled than in the compression-rolled specimens, and as rolling proceeds this may cause greater work-hardening in the straight-rolled specimens.

REFERENCES.

1. W. G. Burgers, "Handbuch der Metallphysik". Band 3, Teil 2. Leipzig: **1941**.
2. W. G. Burgers and J. Sandee, *Physica*, 1942, **9**, 996.
3. W. G. Burgers and J. C. M. Basart, *Z. Physik*, 1929, **54**, 74.
4. O. Dahl and F. Pawlek, *Z. Metallkunde*, 1936, **28**, 266.
5. M. Cook and T. Ll. Richards, *J. Inst. Metals*, 1940, **66**, 1.
6. F. Pawlek, *Z. Metallkunde*, 1935, **27**, 160.
7. G. Wassermann, *Z. Metallkunde*, 1936, **28**, 262.
8. W. G. Burgers and J. J. A. Ploos van Amstel, *Physica*, 1937, **4**, 15.
9. W. G. Burgers and J. J. A. Ploos van Amstel, *Physica*, 1938, **5**, 313.
10. H. G. Müller, *Z. Metallkunde*, 1939, **31**, 161.
11. C. S. Barrett, *Trans. Amer. Inst. Min. Met. Eng.*, 1939, **135**, 296.
12. J. C. Chaston, *J. Inst. Metals*, 1945, **71**, 23.
13. C. S. Barrett and L. H. Levenson, *Trans. Amer. Inst. Min. Met. Eng.*, 1940, **137**, 76.
14. P. Jacquet, *Compt. rend.*, 1935, **201**, 1473.
15. L. I. Gilbertson and O. W. Fortner, *Trans. Electrochem. Soc.*, 1942, **81**, 199.
16. A. de Sy and H. Haemers, *Aluminium*, 1942, **24**, 96.
17. M. Polanyi and G. Sachs, *Z. Metallkunde*, 1925, **17**, 227.
18. M. Bonzel, *Rev. Mét.*, 1937, **34**, 372, 429.
19. W. G. Burgers, *Physica*, 1942, **9**, 987.
20. G. Sachs and E. Schiebold, *Z.V.d.I.*, 1925, **69**, 1557.
21. R. M. Brick and M. A. Williamson, *Trans. Amer. Inst. Min. Met. Eng.*, 1941, **143**, 84.
22. C. S. Barrett, *Trans. Amer. Inst. Min. Met. Eng.*, 1941, **143**, 91 (discussion).
23. C. H. Mathewson and K. R. Van Horn, *Trans. Amer. Inst. Min. Met. Eng.*, *Inst. Metals Div.*, **1930**, 59.

A NOTE ON THE ALUMINIUM–MANGANESE *G* PHASE.* 1133

By (Miss) K. LITTLE,† M.A., B.Sc., Student Member, and
W. HUME-ROTHERY,‡ M.A., D.Sc., F.R.S., Member.

Synopsis.

Previous work has shown the existence of a metastable phase, denoted *G*, in aluminium-rich alloys of aluminium and manganese made from high-purity metals and annealed at low temperatures. Debye–Scherrer films of this phase have been taken and measured, and the lines can be indexed as belonging to a primitive cubic lattice with lattice spacing $a = 13{\cdot}25$ kX. for the side of the unit cell. A survey has been made of the effects of silicon, iron, cobalt, nickel, copper, magnesium, zinc, and chromium on the stability of the *G* phase in Al–Mn alloys containing small quantities of each of the above-quoted metals. Chromium is the only element so far examined which causes *G* to become a stable constituent. An addition of iron or silicon, in comparatively small amounts (about 0·2%), completely suppresses the formation of *G* in alloys annealed at 550° C. It appears improbable that *G* will be found in commercial alloys which contain iron or silicon in more than very small percentages.

I.—Introduction.

Previous work [1] by Little, Raynor, and Hume-Rothery has shown that in binary Al–Mn alloys a metastable phase denoted *G* is found in alloys containing a few per cent. of manganese when these have been annealed at not too high a temperature. The composition of the *G* phase was estimated as being approximately $MnAl_{12}$, and the experimental work suggested that it was never a stable phase of the binary Al–Mn system. The *G* phase was, however, very persistent in alloys annealed at temperatures below about 550° C., and this suggested that the free-energy–composition curve for the phase was only slightly higher than would be required to make it a stable phase of the system. It was, therefore, to be expected that the addition of relatively small amounts of a third element might lead either to the appearance of *G* as a stable phase or, alternatively, to its complete disappearance from ternary alloys. The discussion on this paper [2] showed that the *G* phase was present in alloys prepared in other laboratories, and Marchand [2] claimed that the majority of the lines on Debye–Scherrer films could be indexed as belonging to a body-centred cubic structure with the side of the unit cube $a = 7{\cdot}47$ kX. In the present note the authors

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† British Nylon Spinners, Ltd., Coventry; formerly Inorganic Chemistry Laboratory, Oxford.

‡ Royal Society Warren Research Fellow, Oxford.

describe an examination of the X-ray films, together with some preliminary data regarding the effect of different elements on the stability of the *G* phase in ternary alloys.

II.—X-Ray Diffraction Data.

Measurements were made of the strongest lines due to the *G* phase in binary Al–Mn alloys of the ($\alpha + G$) type, and of the corresponding lines from the ternary compound *G* obtained in an almost pure state

Table I.

Line.	Angle.	*d.*	$h^2+k^2+l^2$.	*a.*	Possible Miller Indices.
1	8·63	5·125	6	12·55	2.1.1
2	14·85	3·001	24	13·10	4.2.2
3	17·07	2·618	25	13·09	5.0.0/4.3.0
4	19·29	2·328	32	13·17	4.4.0
5	19·60	2·292	33	13·17	4.4.1/5.2.2
6	20·05	2·243	34	13·08	5.3.0/4.3.3
7	20·56	2·190	36	13·14	6.0.0/4.4.2
8	21·15	2·131	38	13·14	6.1.1/5.3.2
9	21·60	2·089	40	13·21	6.2.0
10	21·90	2·061	41	13·20	5.4.0/6.2.1/4.4.3
11	22·93	1·974	45	13·24	6.3.0/5.4.2
12	24·56	1·850	51	13·21	7.1.1/5.5.1
13	26·05	1·750	57	13·21	7.2.2/5.4.4
14	28·98	1·587	69	13·19	7.4.2/8.2.1
15	30·48	1·516	76	13·21	6.6.2
16	31·84	1·500	77	13·16	6.5.4/8.3.2
17	33·1	1·408	88	13·20	6.6.4
18	34·4	1·361	94	13·19	9.3.2/7.6.3
19	37·05	1·276	107	13·20	9.5.1/7.7.3
20	37·8	1·257	110	13·19	10.3.1/9.5.2/7.6.5
21	38·27	1·215	118	13·20	9.6.1/10.3.3
22	39·4	1·211	120	13·27	10.4.2
23	42·2	1·145	134	13·25	11.3.2/9.7.2./10.5.3
24	44·36	1·100	144	13·20	12.0.0/8.8.4
25	45·1	1·085	148	13·20	12.2.0
26	45·9	1·071	152	13·20	12.2.2/10.6.4
27	46·78	1·055	157	13·22	11.6.0/12.3.2
28	46·9	1·053	158	13·24	11.6.1/10.7.3
29	47·9	1·036	163	13·23	9.9.1
30	48·5	1·026	166	13·22	9.9.2/11.6.3/9.7.6
31	49·09	1·017	169	13·22	13.0.0/12.5.0/12.4.3
32	59·36	0·894	219	13·22	13.7.1/13.5.5/11.7.7/12.8.1/12.7.4
33	60·74	0·881	226	13·25	15.1.0/12.9.1/9.9.8
34	62·17	0·869	232	13·24	14.6.0
35	65·19	0·847	244	13·23	12.10.0/12.8.6
36	70·35	0·816	262	13·24	15.6.1/10.9.9
37	72·31	0·807	269	13·24	13.10.0/16.3.2/12.11.2/14.8.3/ 12.10.5/13.8:6
38	74·48	0·798	275	13·23	15.7.1/15.5.5/13.9.5
39	76·98	0·789	281	13·23	16.5.0/14.9.2/16.4.3/10.10.9/ 14.7.6/12.11.4

from ternary Al–Cr–Mn alloys. The results showed that the substitution of chromium for manganese does not alter the lattice spacing of *G*, and the film for the ternary alloy was therefore used for the purpose of indexing the lines. The results showed clearly that some of the lines cannot be ascribed to the body-centred cubic cell proposed by Marchand. From the data in Table I it will be seen that all the lines can be indexed as belonging to a primitive cubic lattice. The usual extrapolation methods indicate a value $a = 13{\cdot}25$ kX. for the side of the unit cell. It has not been possible to measure the density of the *G* phase, but if the composition is assumed to be $MnAl_{12}$, he known densities of aluminium and manganese suggest that the nsity of the *G* phase will be approximately 3·0, and it seems probable , the unit cell of the *G* phase is associated with 12 molecules of $MnAl_{12}$.

III.—The Effect of Different Elements on the Stability of the *G* Phase.

A preliminary survey has been made of the effects of silicon, iron, cobalt, nickel, copper, magnesium, and zinc on the stability of the *G* phase, while the effect of chromium has already been examined in detail.[3] For this purpose, a few ternary alloys were prepared for each system so as to obtain a section across the aluminium corner of the diagram at 98% aluminium. All specimens were annealed for 8 days at 550° C., and were then examined microscopically. The results obtained may be summarized as follows:

(1) Chromium is the only element so far examined which causes *G* to become a stable constituent.

(2) The addition of zinc to the binary Al–Mn alloys has little effect on the relative proportion of *G* and $MnAl_6$.

(3) The addition of small amounts of copper reduces the proportion of *G* slightly, while larger amounts produce a greater reduction. It has been reported to the authors by A. T. Little[4] that at lower temperatures (400° C.) *G* may be a very persistent phase, while Day and Phillips[5] have also found *G* in Al–Cu–Mn alloys.

(4) The addition of cobalt suppresses *G* considerably, and the relative proportion of *G* to $MnAl_6$ decreases steadily as the percentage of cobalt is increased.

(5) The addition of magnesium also causes a decrease in the stability of *G*, the effect being about the same as that of a corresponding weight percentage of cobalt.

(6) In the system Al–Mn–Ni, the existence of ternary phases

results in $MnAl_6$ or *G* being found only in alloys whose compositions lie near the manganese axis of the ternary diagram. Over this narrow range, the proportions of *G* appear to be less than in the corresponding cobalt alloys.

(7) The addition of approximately 0·2% of iron or silicon completely suppresses the formation of *G* in alloys annealed at 550° C. In the case of iron the effect is very marked, and small percentages of iron will neutralize the effect of chromium. It has also been noted by Axon and Hume-Rothery [6] that the *G* phase is not observed in a wide range of ternary, quaternary, and quinary alloys of the syste Al–Mg–Mn–Zn–Si containing 1% silicon and annealed to equilibri at 460° C.

IV.—Discussion.

From this preliminary survey it appears improbable that the *G* phase will be present in any aluminium alloys which contain silicon or iron in more than very small proportions. Its presence may, however, be expected in Al–Mn alloys made from pure aluminium, and in more complex alloys containing chromium, copper, magnesium, or zinc, provided that iron and silicon are absent. The above-quoted results would also lead one to expect the presence of vanadium to stabilize the *G* phase. The present authors had hoped to examine the whole problem in detail, but circumstances have arisen which have made this impossible, and the present short note is therefore submitted in the hope that the preliminary results may be of interest.

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References.

1. K. Little, G. V. Raynor, and W. Hume-Rothery, *J. Inst. Metals*, 1946, **73**, 83.
2. *J. Inst. Metals*, 1947, **73**, 747.
3. G. V. Raynor and K. Little, *J. Inst. Metals*, 1945, **71**, 493.
4. Private communication.
5. M. K. B. Day and H. W. L. Phillips, *J. Inst. Metals*, 1947, **74**, 33.
6. H. J. Axon and W. Hume-Rothery, *J. Inst. Metals*, 1948, **74**, 315.